SEPTEMBER 1960
Volume 196 Part 1

JOURNAL OF THE IRON AND STEEL INSTITUTE

Established 1869
Incorporated by Royal Charter 1899

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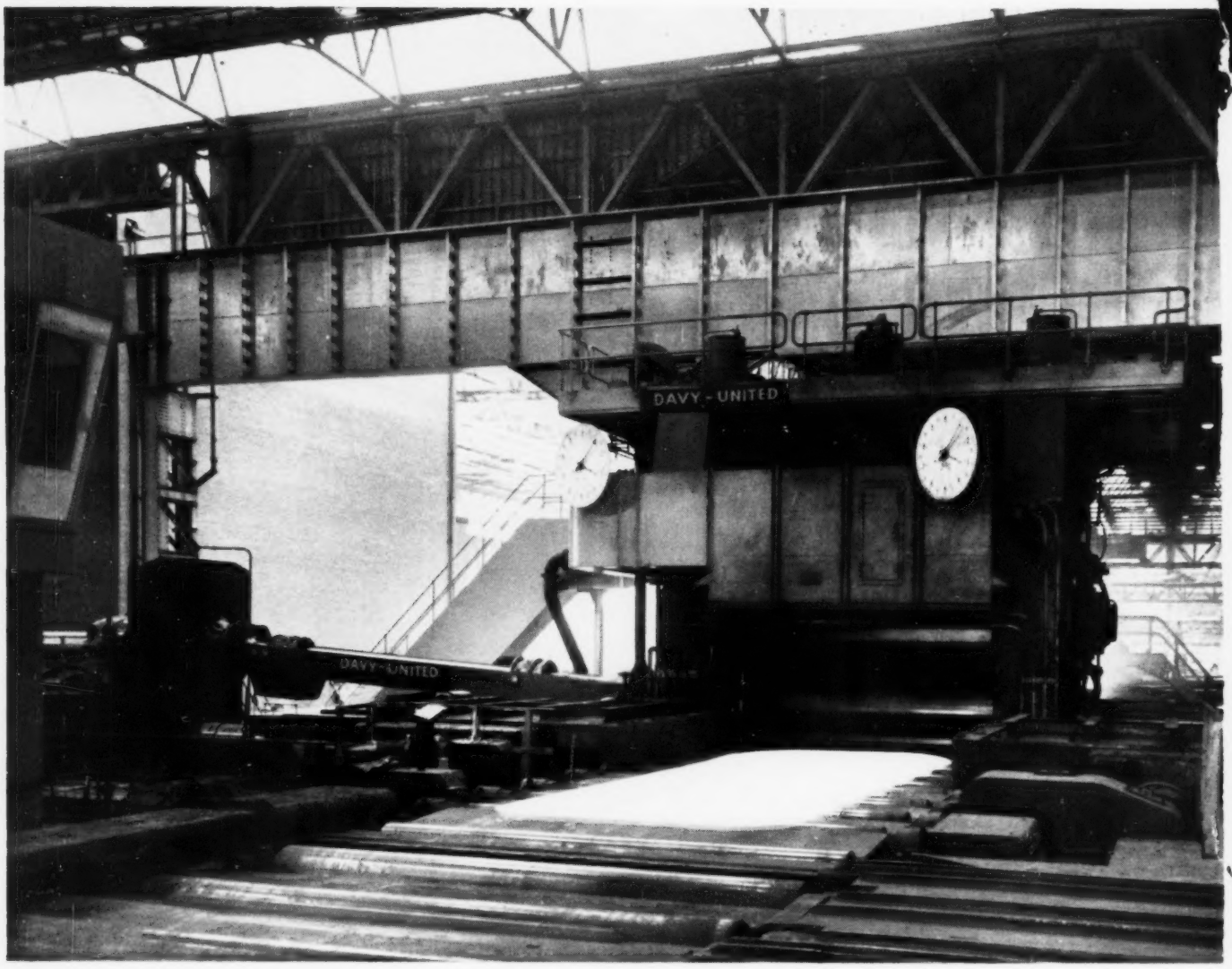
CONTRIBUTION
TO AN INDUSTRY



This new Hot Plate Finishing Mill is now in operation in the works of Appleby-Frodingham Steel Company. Driven by two 4,000 h.p. motors, it can roll high quality steel plate up to 12 feet wide. It is one of three new plate mill plants ordered from Davy-United under current steel industry development plans.

DAVY-UNITED

Issued by the Machinery Division of
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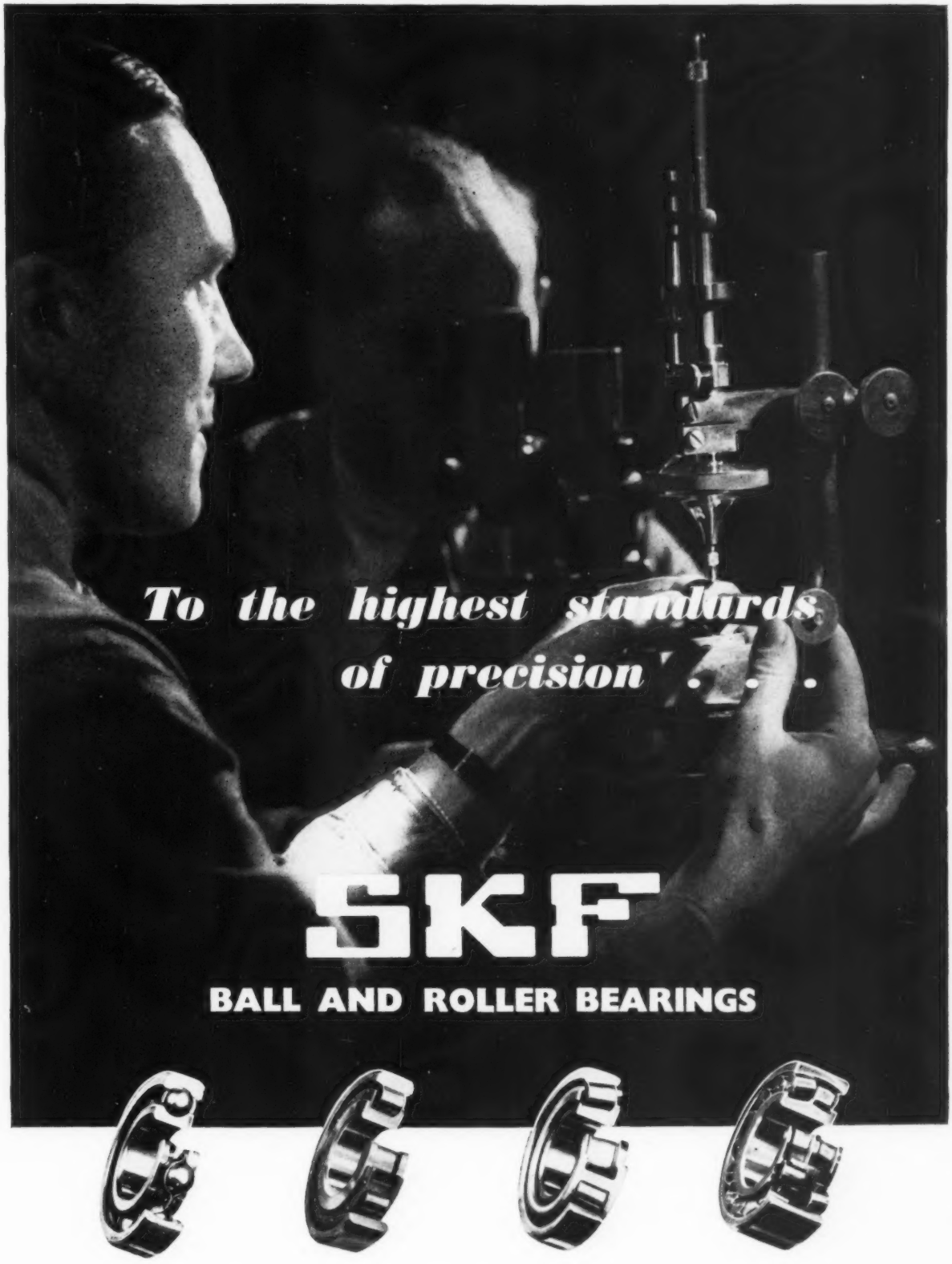
339 DU



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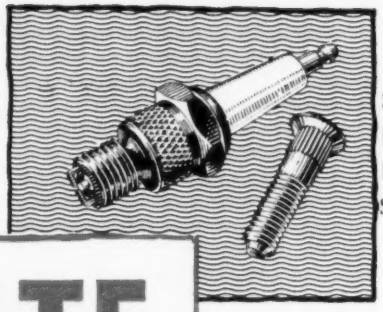

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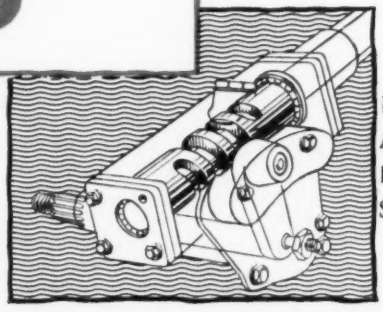

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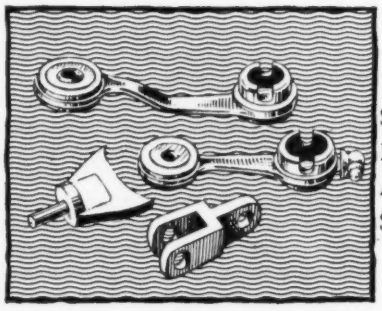

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Coils may be split if required.

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0.550″ to 1.000″ in 900 lb. coils.
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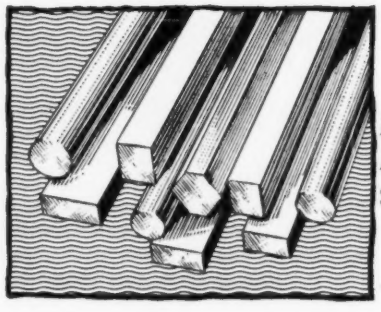

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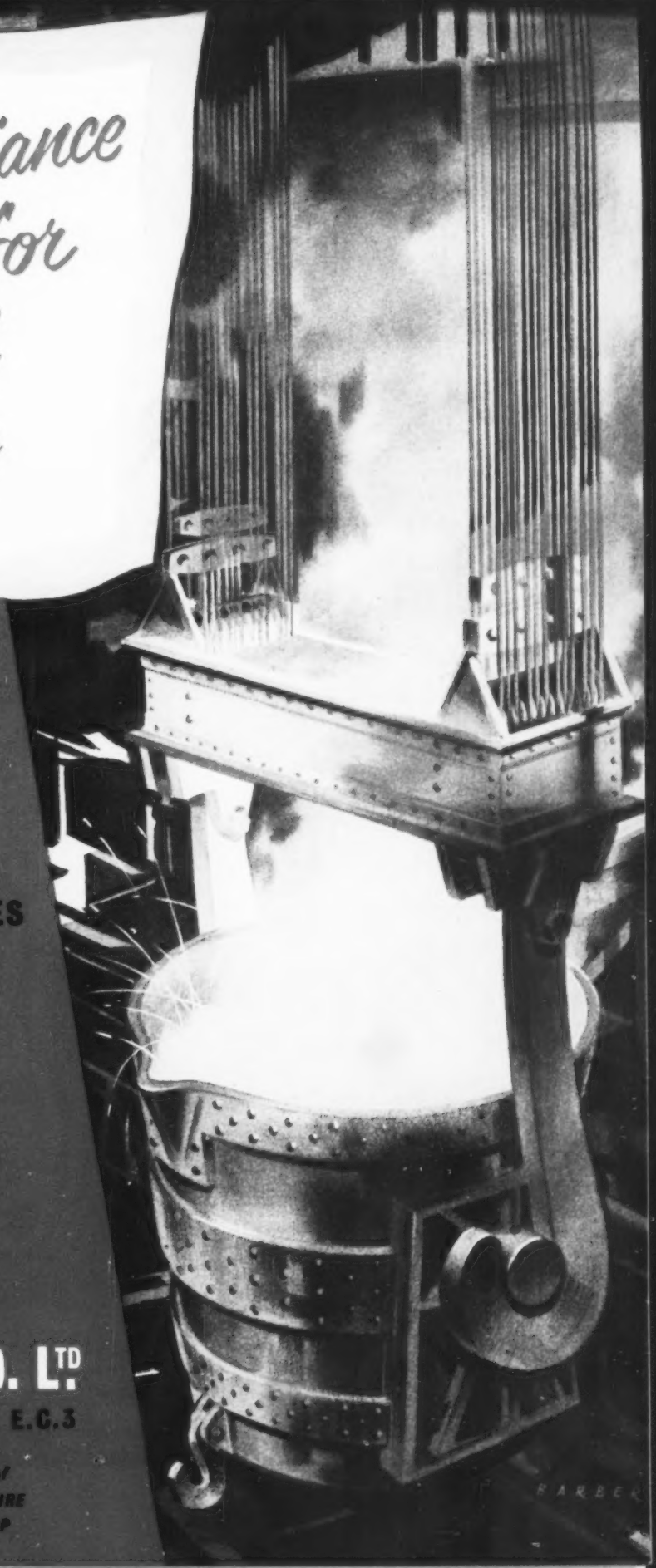

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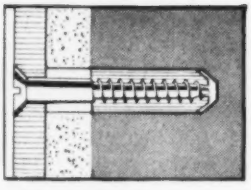
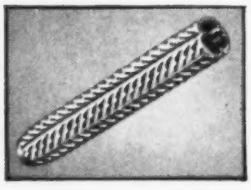
RAWLPLUG FIXING DEVICES and TOOLS . . .

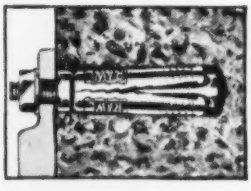
Backed by the Name that Carries the Weight



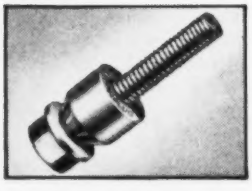
For firm Screw Fixings. The specially treated fibrous construction of the Rawlplug provides the most effective stronghold for wood-screws in any masonry. The simple method of making the hole, inserting the Rawlplug and driving home the screw enables firm fixings to be made quickly without mess or damage to walls or decorations. Sizes are from the tiny No. 3 to the 1″ diameter No. 30.



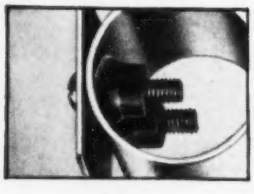
On the occasions where a metal plug is preferred or advised, such as for wet or acid situations, there is the choice of Rawlplug Screw Anchors or White Bronze Plugs, but for these Cadmium plated screws should be used.



Heavy Duty Fixings. RAWLBOLTS are expansion bolts which only require a hole in the masonry in the same way as a Rawlplug. By turning the bolt the metal segments are locked within the hole and the fixing will take very heavy loads with complete safety. Sizes are from 3/16″ to 1″ diameter.

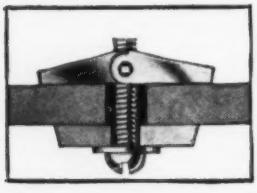


Wet or Acid Situations. Rawlplug Bolt Anchors and Rawltamps are made for heavy duty bolt fixings in places where the exposure to corrosion is extremely high. The Bolt Anchor can be set deep down in thick concrete whereas the Rawltamp enables a threaded insert to be fixed in shallow concrete.

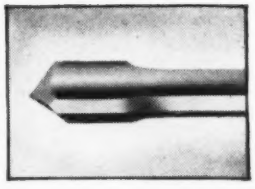


Thin or Cavity Walls. Proprietary building materials often present fixing problems because of being either hard and thin or thick and soft. The Rawlplug Company have devised several clever devices for making firm fixings to such materials including lath and plaster ceilings, hollow pot, panel doors, etc.

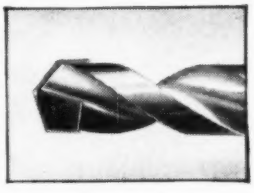
The upper illustration shows how one of the devices will make the almost impossible fixing of a metal plate to a pipe and the bottom illustration shows how the wings of a Spring Toggle spread the load over a plasterboard ceiling.



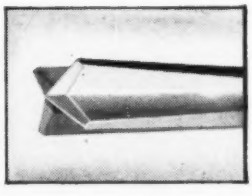
Rawldrills. Standard sizes for Rawlplugs Nos. 3 to 30 and Rawlbolts A. B. C. D. E. and G. Adaptable Rawldrills need only three holders for the 10 sizes Nos. 3 to 22. There is also a very useful Universal Tool-set with a knurled holder to take Universal Rawldrills Nos. 6 to 20.



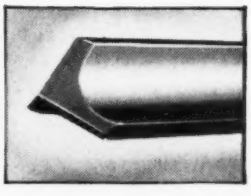
Durium tipped Drills. For faster, silent penetration of masonry and tiles Durium drills are the answer. They can be used in hand or suitable electric drills. Sizes for Rawlplugs Nos. 6 to 30; for C. D. E. and G. Rawlbolts, and a long series for drilling right through walls. Free resharpening Voucher given with every Durium drill.



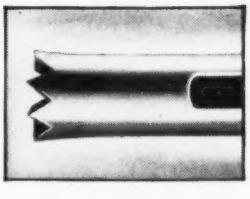
Rawlplug Stardrills. An inexpensive one piece tool for use on jobs needing a small number of holes. Twenty sizes from 11/32″ to 2½″ are made, of which eight are for Rawlbolts "A" to "K", and it is only necessary to quote the reference letters of the Rawlbolts when ordering Stardrills to use with them.



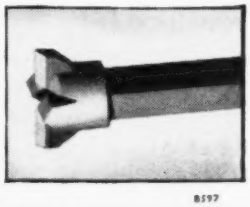
Wall Boring Tools. Specially designed for rapid boring right through walls. This triple fluted percussion tool with hexagon handle in one complete unit has been proved to make a ¾″ hole right through a 9″ stock brick wall in nine minutes. Lengths 18″ and 24″, diameters from 5/16″ to 1½″.



Tubular Boring Tools. Will bore a clean hole in soft brick wall quicker than any other hand percussion tool. The serrated edges saw through masonry and the dust is collected in the channel and ejected through the elongated slot. When working in deep holes the tool should be withdrawn periodically and the dust tapped out. Sizes are from ¼″ to 1″ diameter.



Power Tools. In addition to the hand tools listed above special Rawlplug tools are made for use in electric and pneumatic power tools. Details of these can be had on application.



Rawlplug Fixing Devices and Hole Boring Tools are sold by Ironmongers, Hardware Dealers, Builders Merchants and Stores. If you experience any difficulty in obtaining the type and size you require please send details to the following address.

B597

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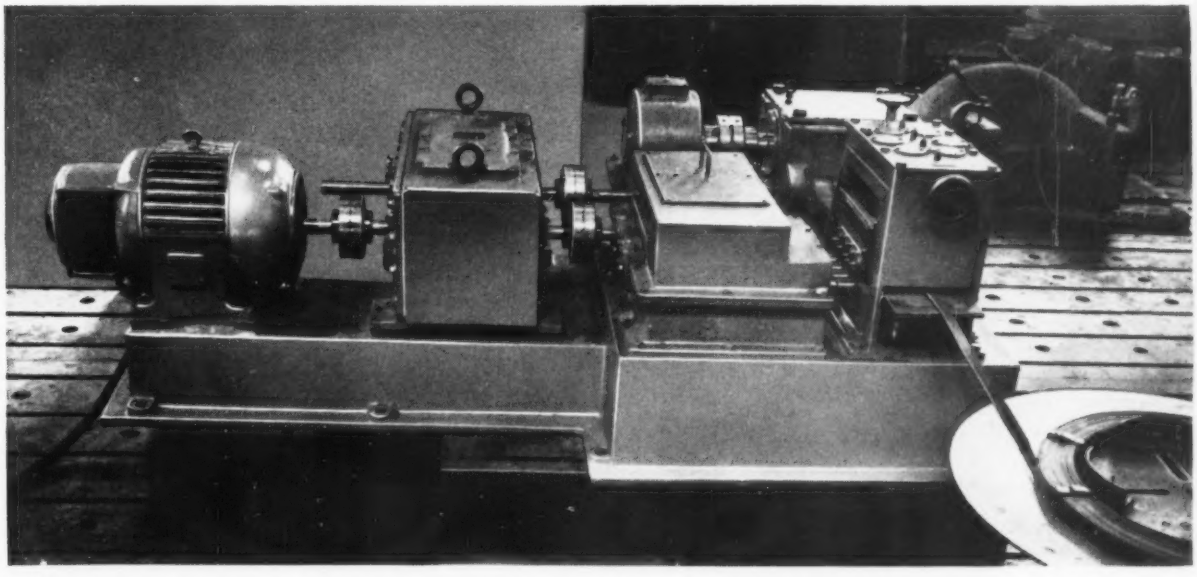

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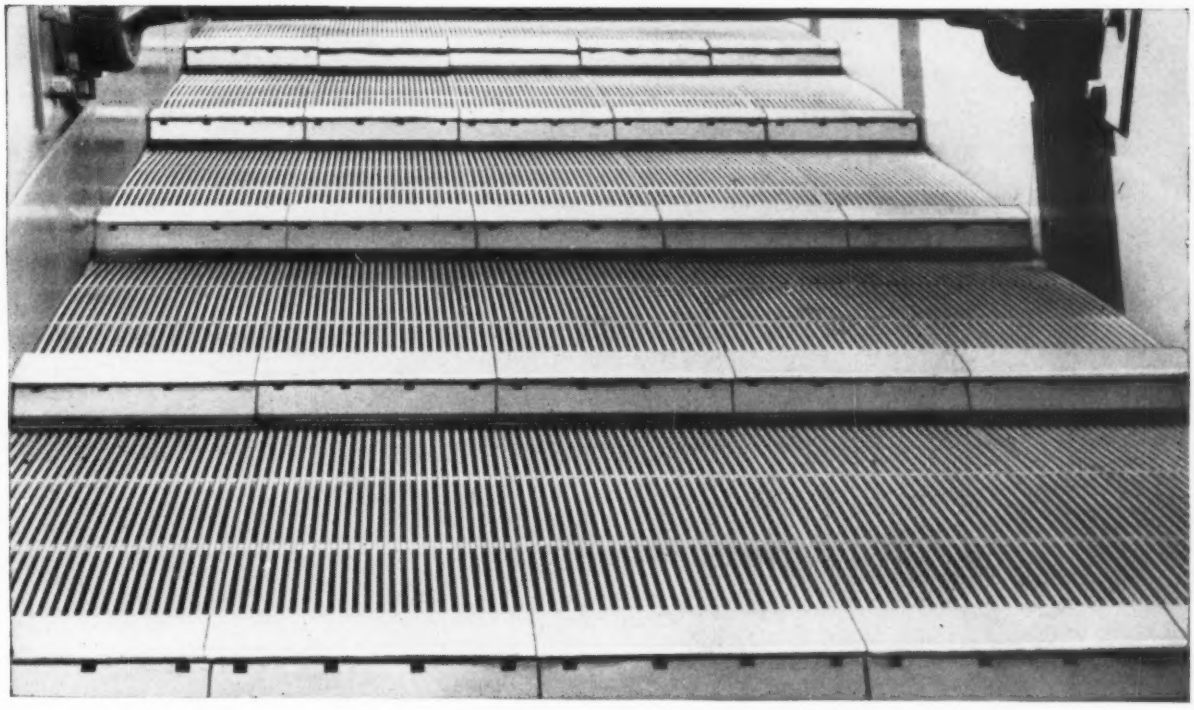

Illustration shows a typical Ty-rock Rod Deck Screen

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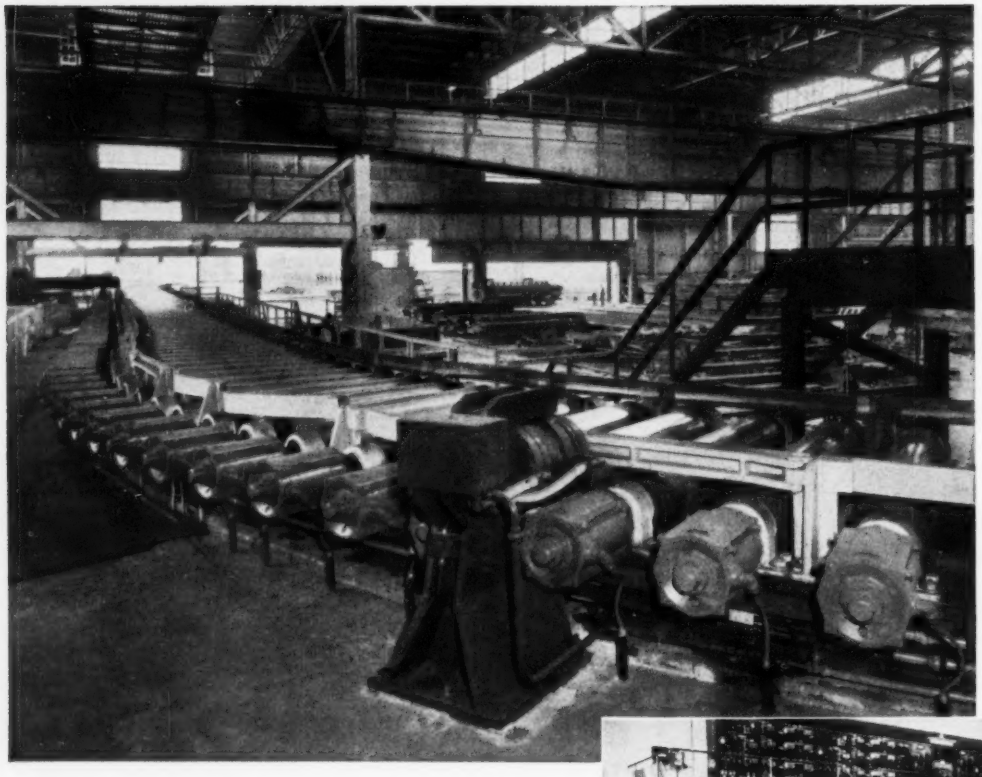

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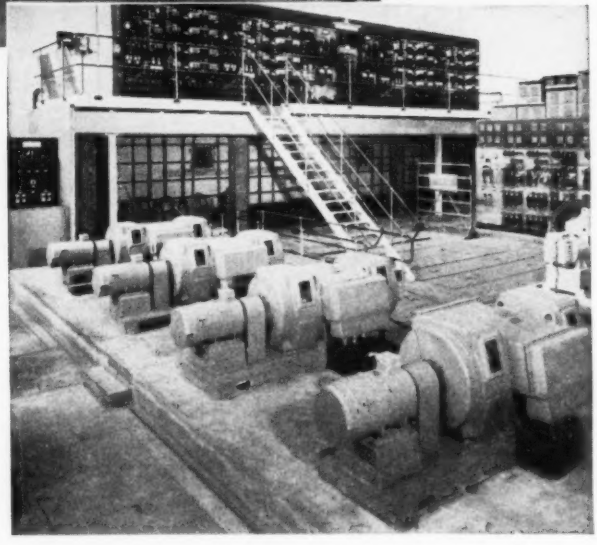

Some of the 210 floating-type motors, and the N-S frequency convertor equipment, providing 36 r.p.m. gearless drive for slab transfer table rolls, at the Abbey Works of the Steel Company of Wales Ltd.

Details of the L.S.E. floating-type motors are given in Publication 113/2, available on request to Laurence, Scott & Electromotors Ltd., Publicity Dept., 376 Strand, London W.C.2

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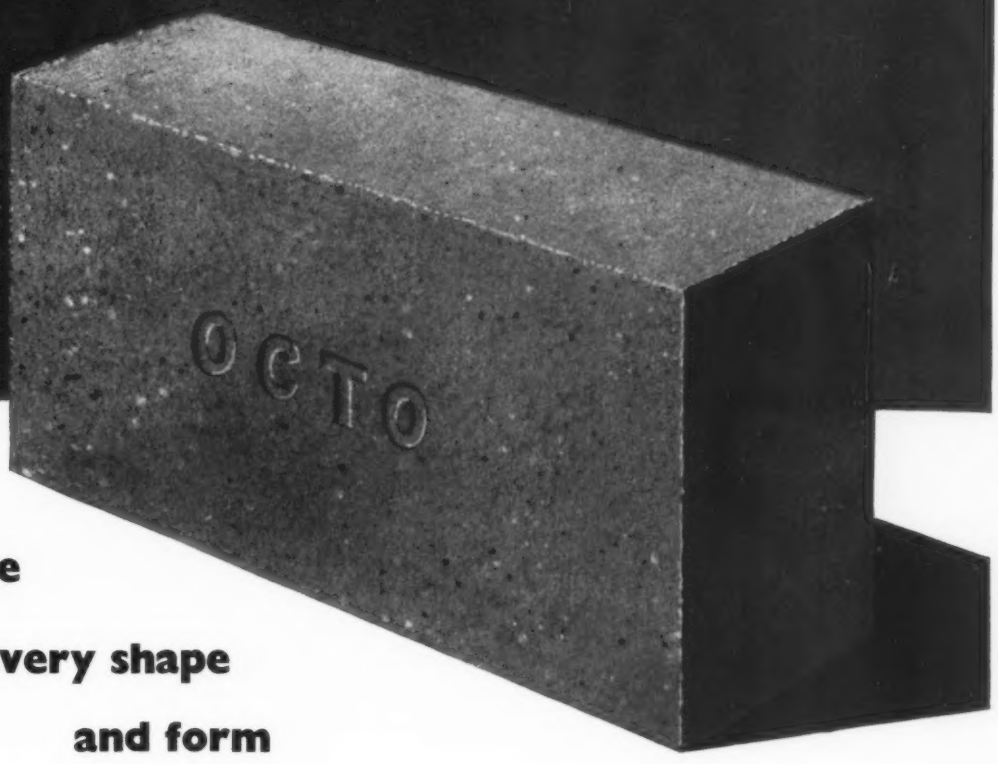

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- Relieves strain on the operator
- Makes operation safer

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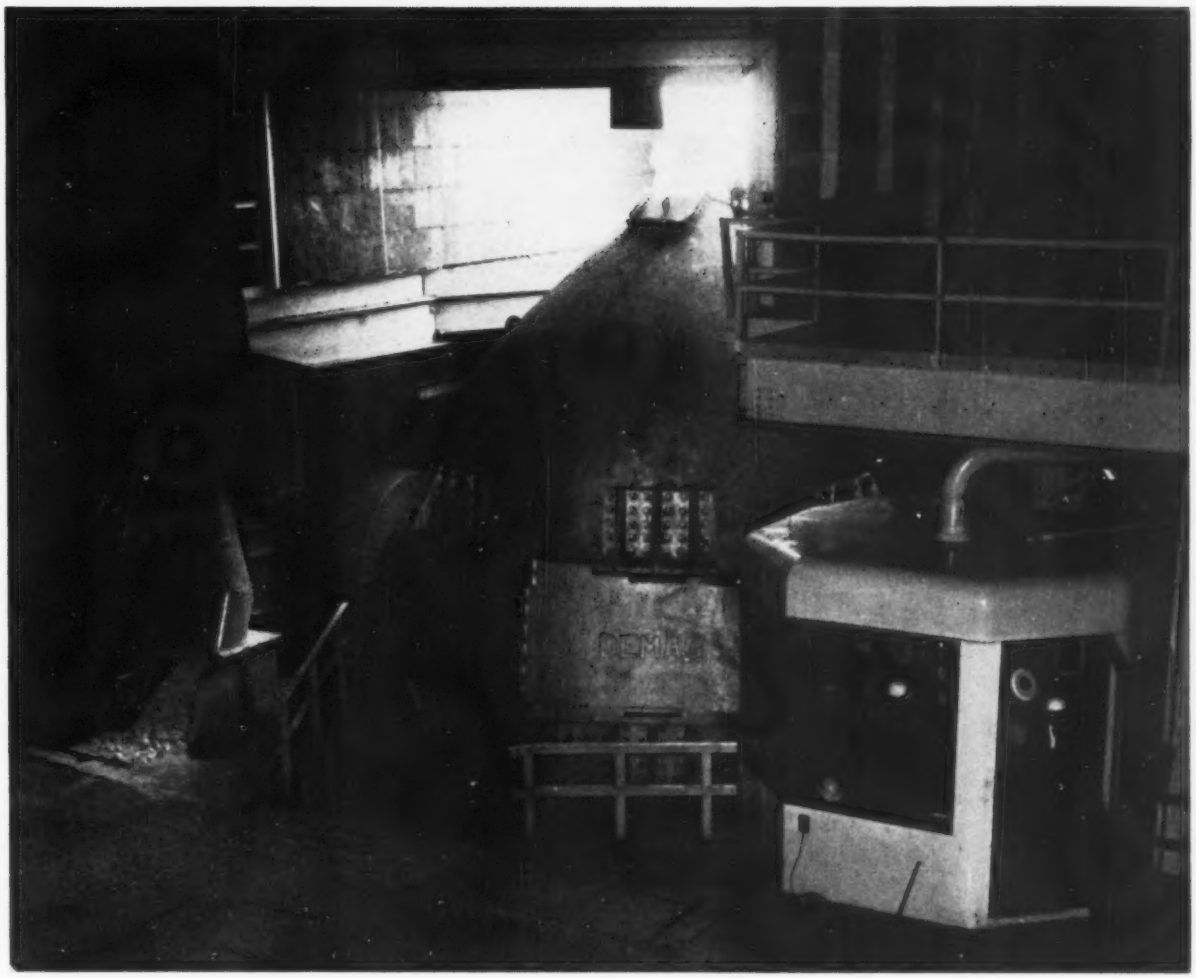

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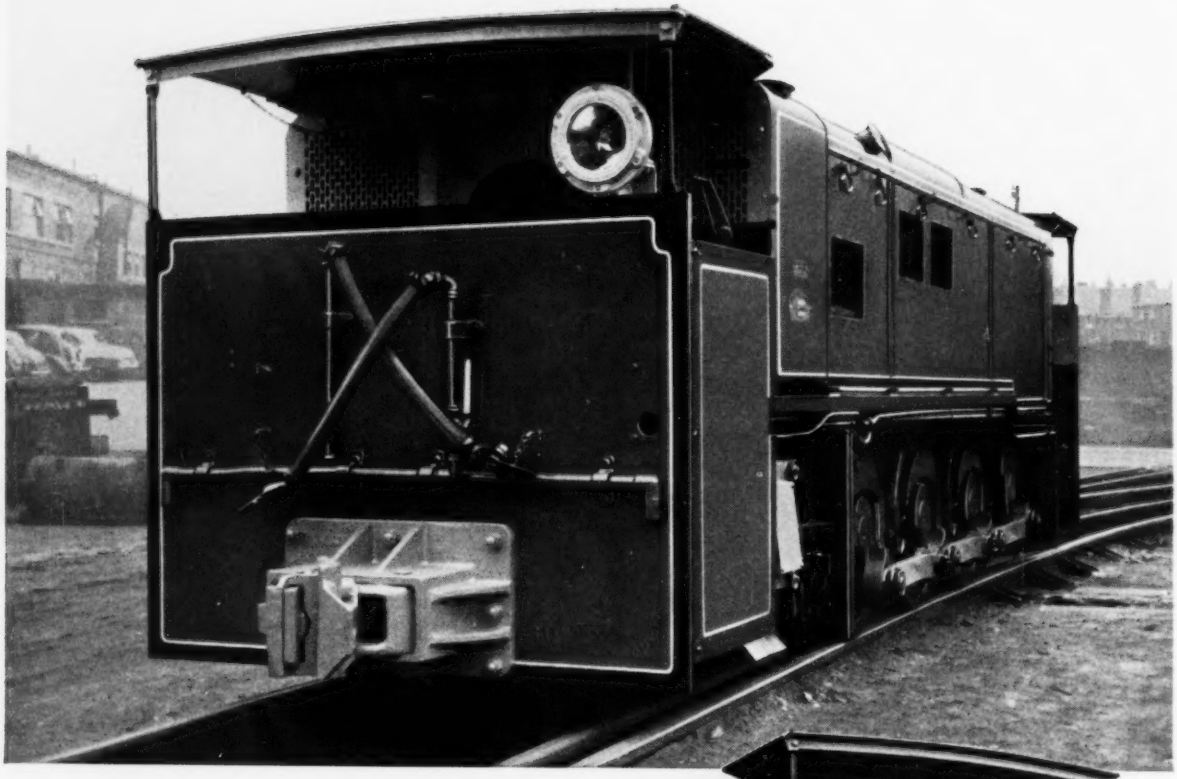

100 H.P., 0-6-0 type, 3′ 6″ gauge diesel locomotive, maximum height 5′ 6″ supplied to Messrs. Stewart & Lloyds Minerals Limited, Thistleton Mine.
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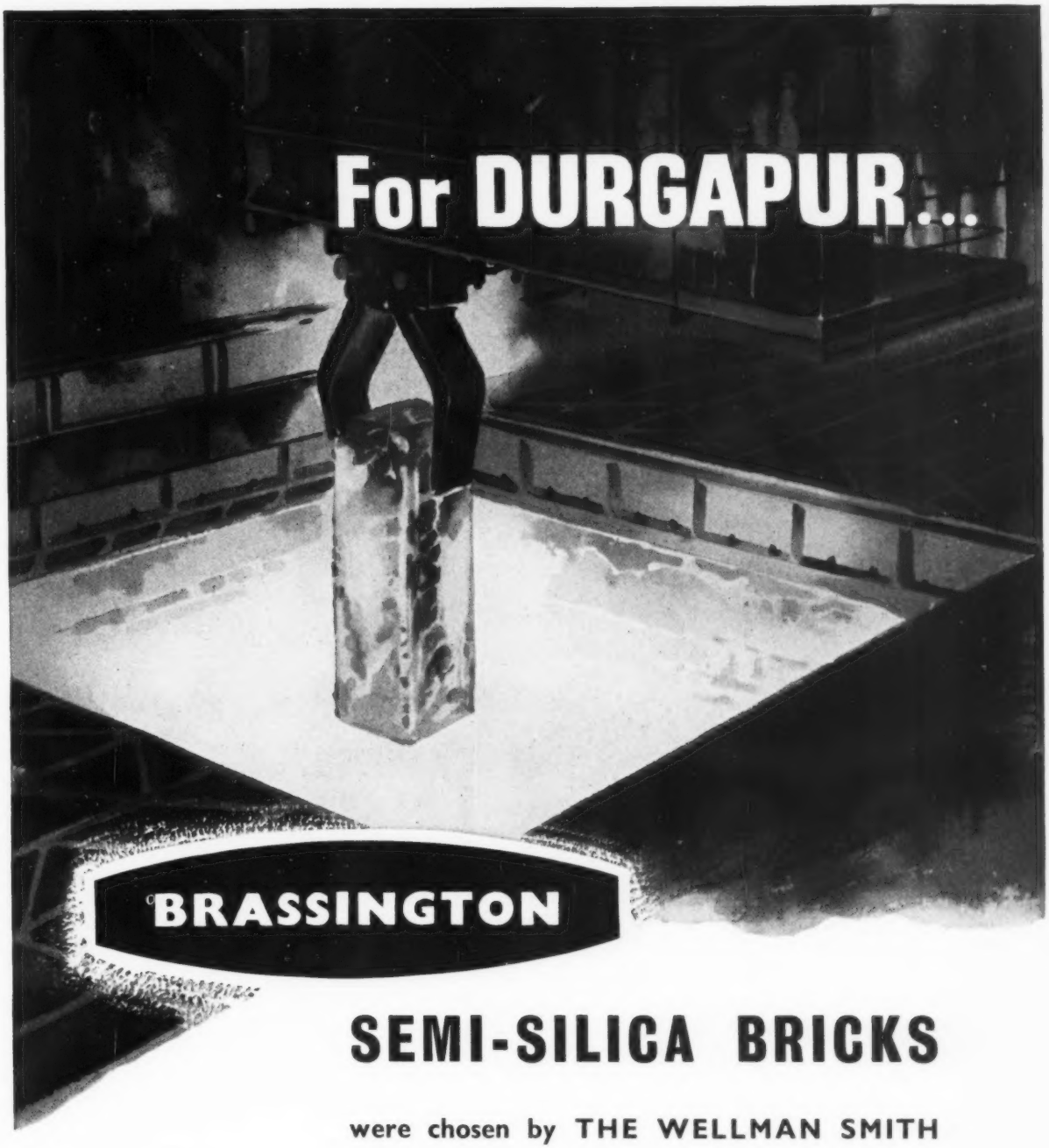
H.P. 7585



SEMI-SILICA BRICKS

were chosen by THE WELLMAN SMITH OWEN ENGINEERING CORPORATION LTD for lining their Soaking Pits at Durgapur

SWANN, RATCLIFFE & COMPANY (BRASSINGTON) LIMITED

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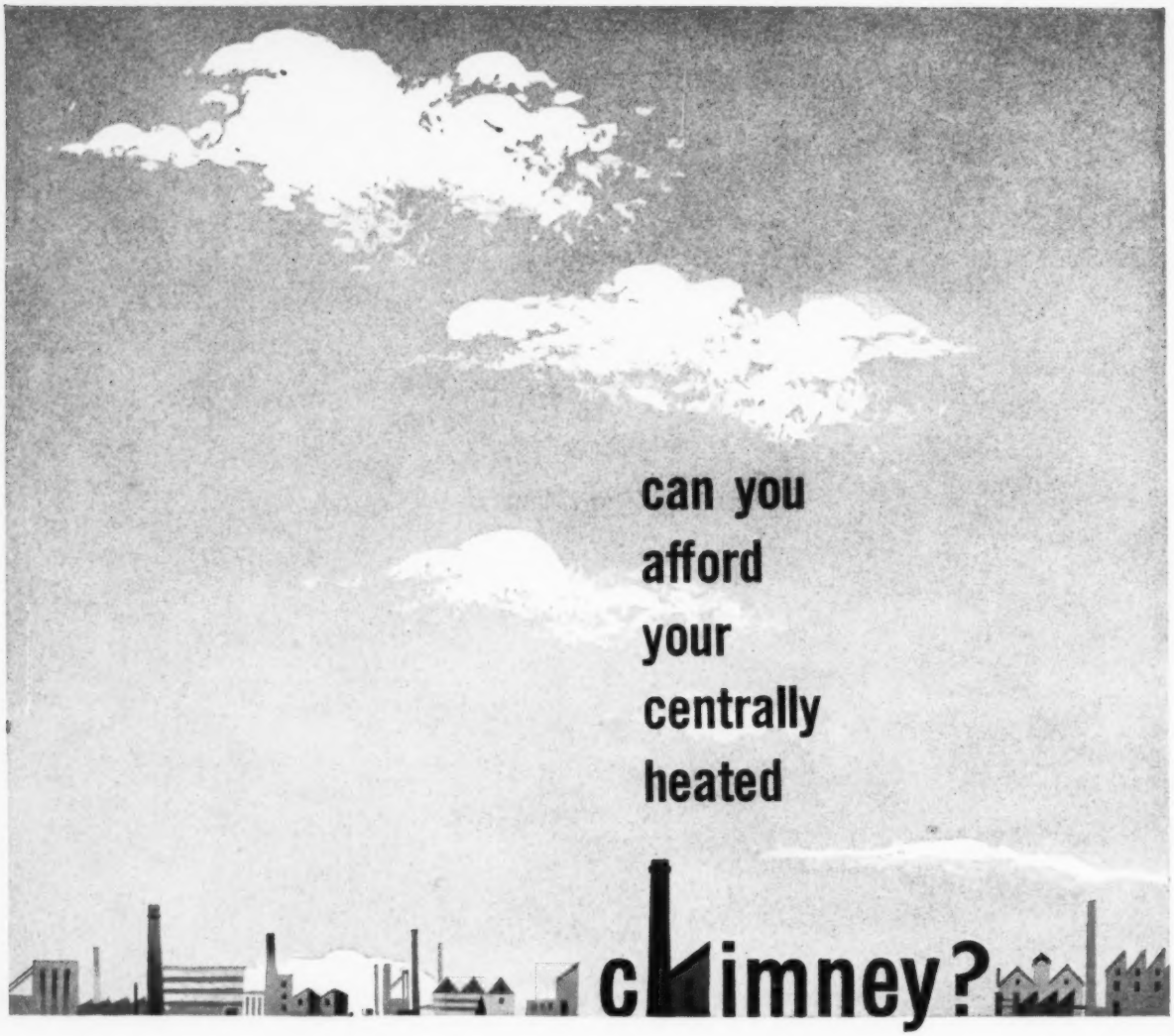

With rising production costs it's hardly likely. The people to see are Spencer-Bonecourt-Clarkson who for over forty years have been experts in waste heat recovery for steam raising in a wide range of industries. In addition to fire-tube boilers, *S-B-C now also supply water-tube boilers.*

The vigorous approach of S-B-C technicians leads them into many fields of enquiry. They are experts on latent cooling techniques as applied to door frames and other parts of open hearth furnaces, and to the skids of reheating furnaces. Keeping abreast of the latest developments in the Steel and other industries, a team of S-B-C engineers recently toured Germany and other continental countries to inspect the newest techniques.

S-B-C are the people best equipped to advise you and to act for you in the recovery of waste heat. Write or 'phone today — time spent now can save you pounds for years to come.

SBC **SPENCER-BONECOURT-CLARKSON LTD** *Leaders in Waste Heat Recovery*
A subsidiary company of Babcock & Wilcox Ltd.
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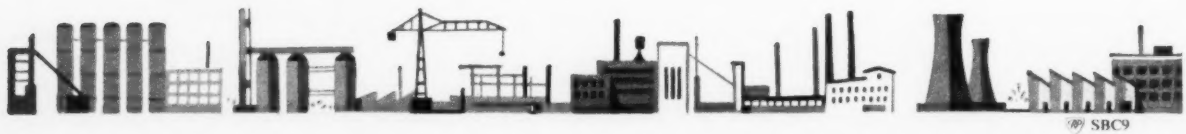

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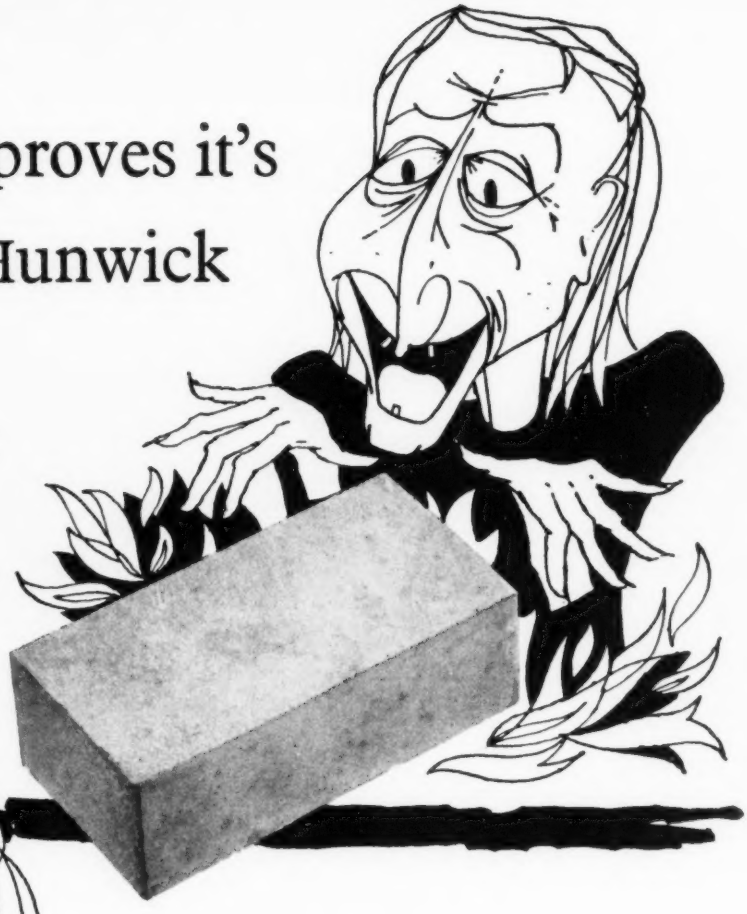
. . . witch proves it's West Hunwick

thermal spalling

send brick falling—

flame impingement

sear and singe it



Accursed Silicas!...

. . . every imprecation

meets frustration—

this must be another brick

*incorruptible, inviolable—*WEST HUNWICK

HUNNEX Brand—high duty quality
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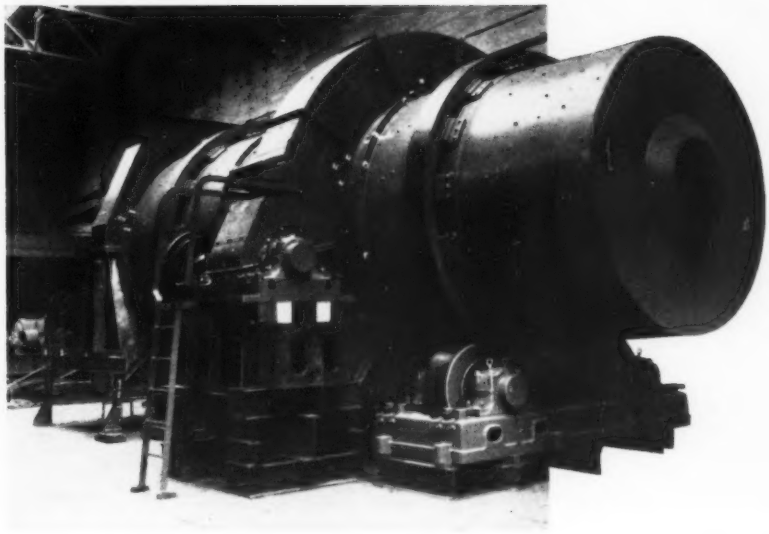
Newells

RAW MATERIAL PREPARATION PLANTS FOR STEELWORKS.



● Another feature of Newells designs.

Rotary Conical Conditioning Drum for Sinter Return Fines.

These photographs illustrate a machine erected in our Works prior to despatch.

This machine is now installed at the Works of Messrs. John Lysaght's Scunthorpe Works Limited, the main contractors being Messrs. Head Wrightson Iron and Steel Works Engineering Limited.

This machine as illustrated is designed for a capacity of 250 tons per hour.

Our designs include machines having capacities up to 450 tons per hour and all sizes up to 15'0" diameter x 38'0" long.

● Other specialities includes :-

Patented Rod Mill Pulverisers for grinding Coke Breeze and Limestone.

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Heavy Duty Table Feeders up to 12'0" diameter.

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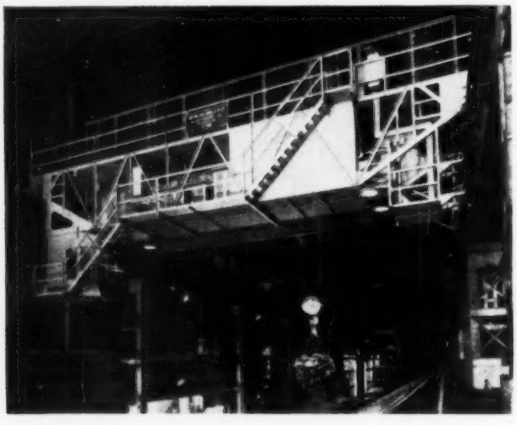


Steelworks Plant

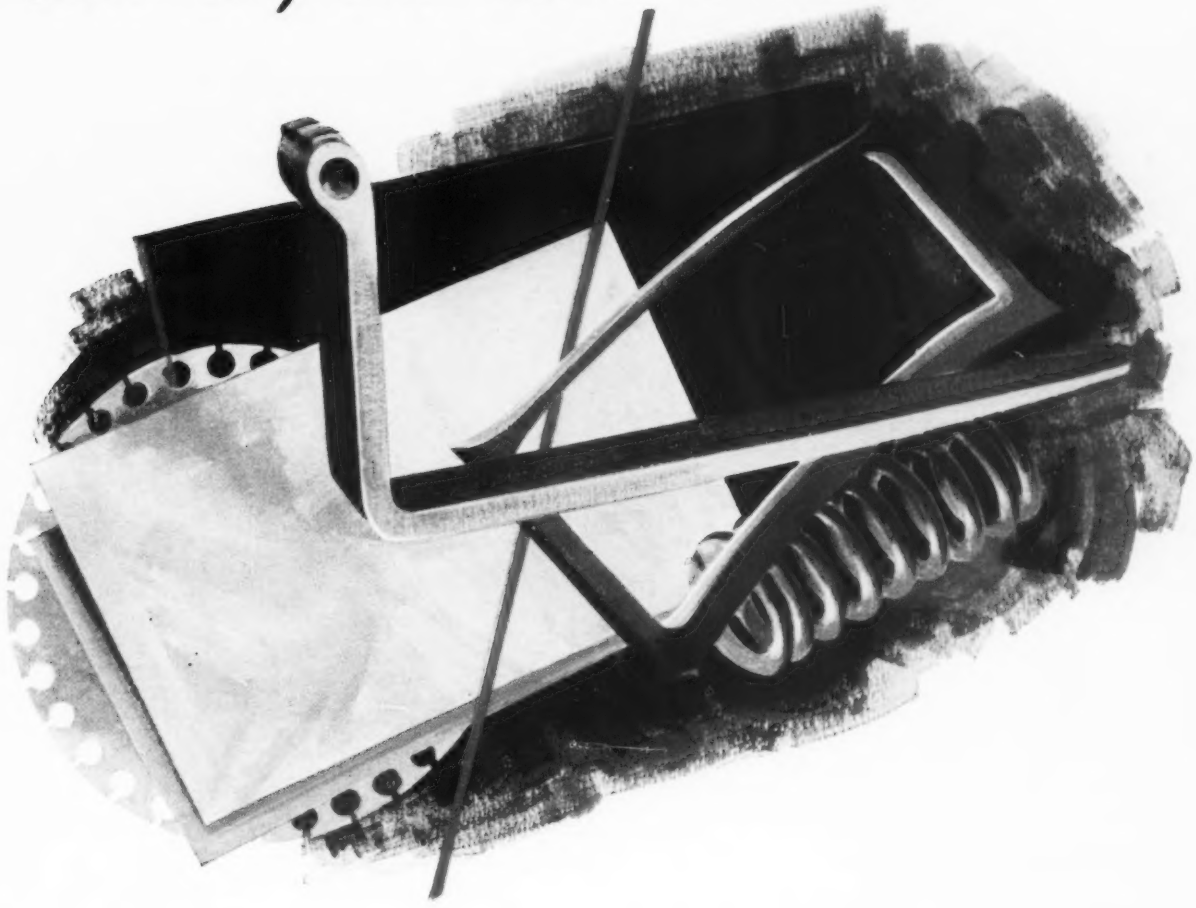
This 10 ton Magnet Crane and the 7 ton overhead Charging Machine illustrated above—one of four such machines—are recent examples of a large quantity of plant supplied to a major steelworks. ARROL are similarly associated with a number of steel companies and their nameplate can be found in most British and many overseas works.

All types of Cranes for Steel Works, Forges, Foundries; Transporters and Ore Bridges; Bunkers and Storage Bins; Steel-framed Buildings and other steel structures, Mechanical Engineering work; Charging machines; H.P. Hydraulic Pumps; Hydraulic Presses and Motors.

SIR WILLIAM ARROL & CO LTD GLASGOW

Arrol

When they talk about MOTORING—



think of R·T·B

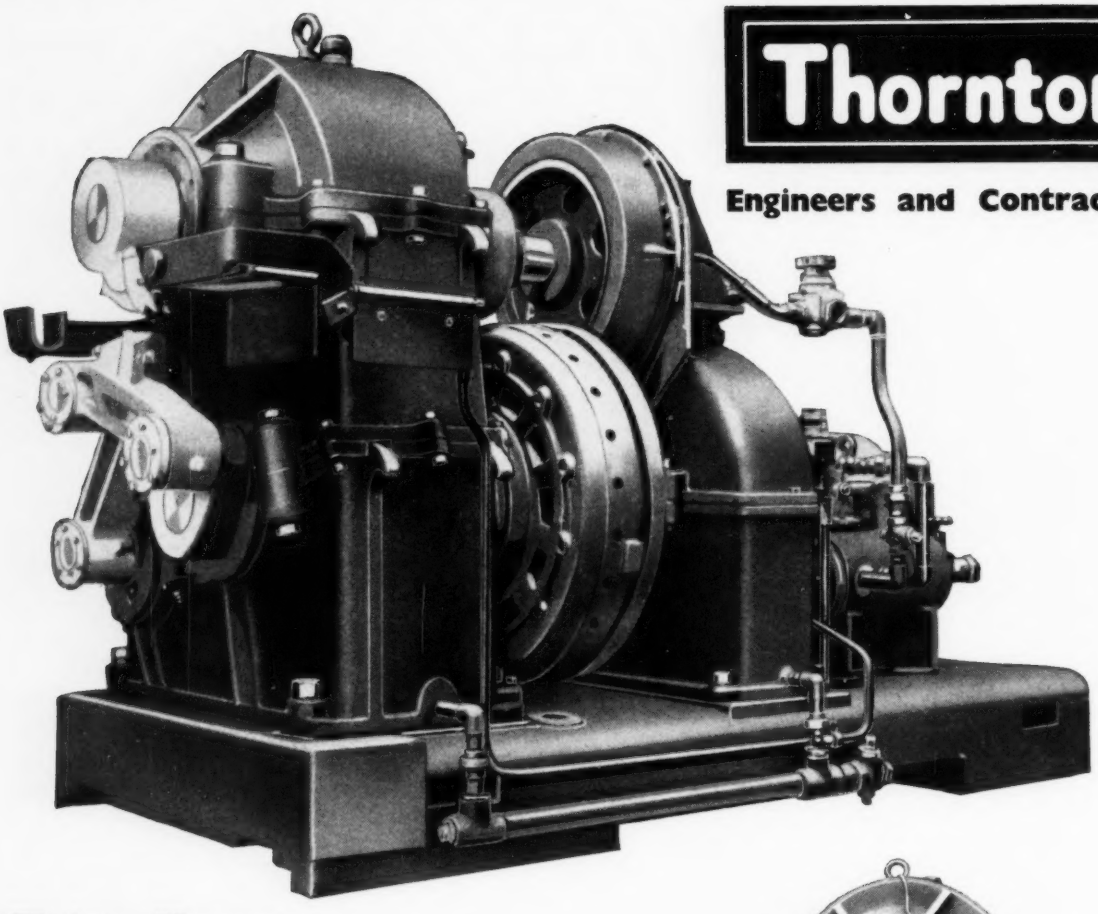
The artist's sketch shows finished parts made from RTB steel— against some steel plate, autobody sheet, and ternecoated sheet.

The steel frames of their car seats are probably pressed from RTB sheet, so are the bodies; the springs of the independent suspension may be made of our alloy steel; so are the 'strong arms' of the fork-lift trucks used by the motor-car manufacturer.

And at home they are served by tinplate, bright sheets and black steel sheets. Hardly a household has less than a hundred examples; and not far away are RTB laminations, generating electricity or turning it into power.

Richard Thomas & Baldwins Ltd

R·T·B



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Bogies of all kinds for carrying light or heavy loads.
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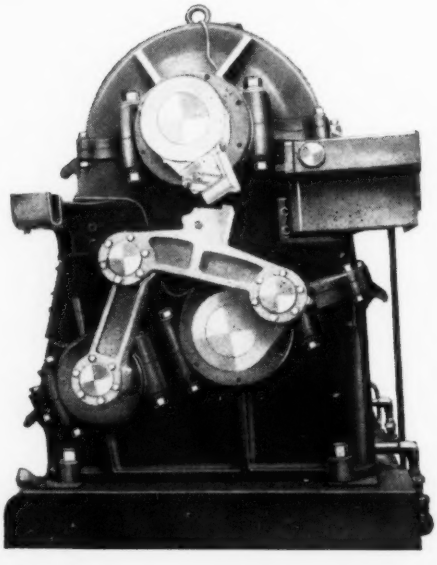
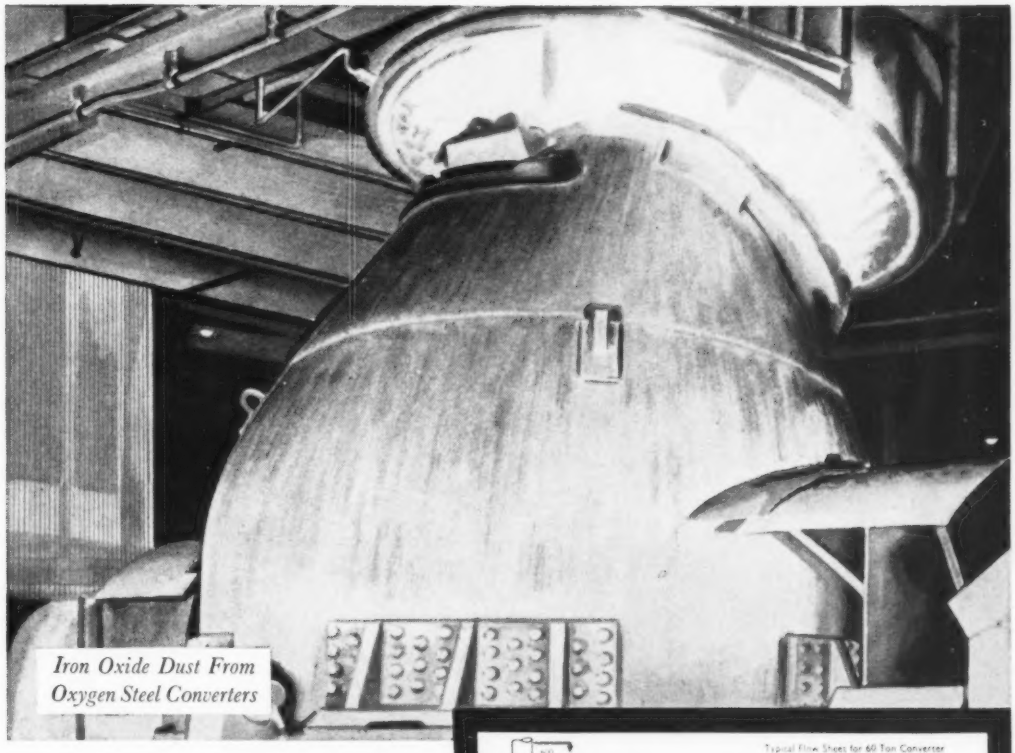
We shall be glad to quote against your requirements and specifications.



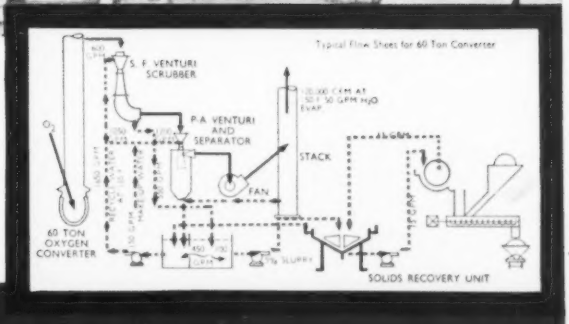
B. THORNTON LIMITED

Turnbridge, Huddersfield. Phone. Huddersfield 7541



Iron Oxide Dust From Oxygen Steel Converters

remove it wet... recover it dry



The recent installation of Chemico Venturi Gas Scrubbers at a major steel plant in Canada marks the sixth time Venturi Gas Scrubbers have been used to prevent an extremely undesirable source of air pollution —Iron Oxide Dust. Chemico Venturi Scrubbers maintain leadership in the world's steel industries because they offer wet scrubbing which allows non-explosive collection, efficiencies above 99% and extremely low maintenance. In addition no critical controls are required to maintain efficiency and the handling of dust is simplified by recovery in a dry form. Chemico Gas Scrubbers are ideally suited for Open Hearth Furnaces, Cupolas, Scarfing operations, Sintering operations, and Electric Furnaces. For further information and a copy of a brochure giving full particulars, contact the Chemico Gas Scrubber Division.

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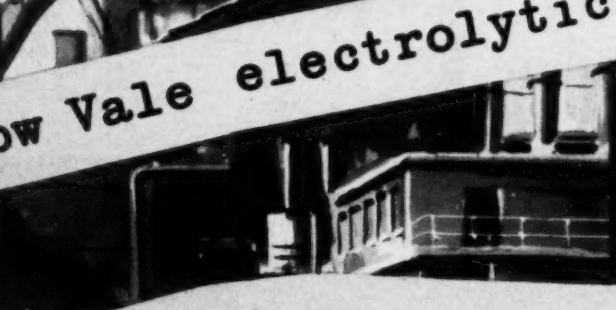
A SUBSIDIARY OF CHEMICAL CONSTRUCTION CORPORATION, NEW YORK, U.S.A.

:-recent world-wide contracts include :-

BEHIND

:-:-durgapur blast furnace plant £6,500,000 :-

:-:-:-ebbw Vale electrolytic tinning plant



:-:-:-poland hot dip tinning line £180,000:-

aviles blast furnace and sinter plant £6,000,000:-

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THE STEEL INDUSTRY WITH

HEAD WRIGHTSON

£2,500,000

TEESDALE IRONWORKS THORNABY-ON-TEES

TEESDALE IRONWORKS THORNABY-ON-TEES

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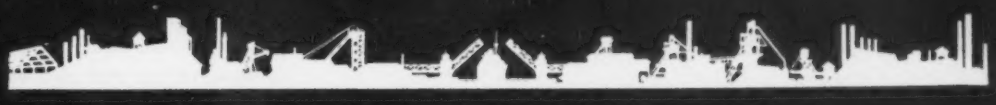

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and PARSONS

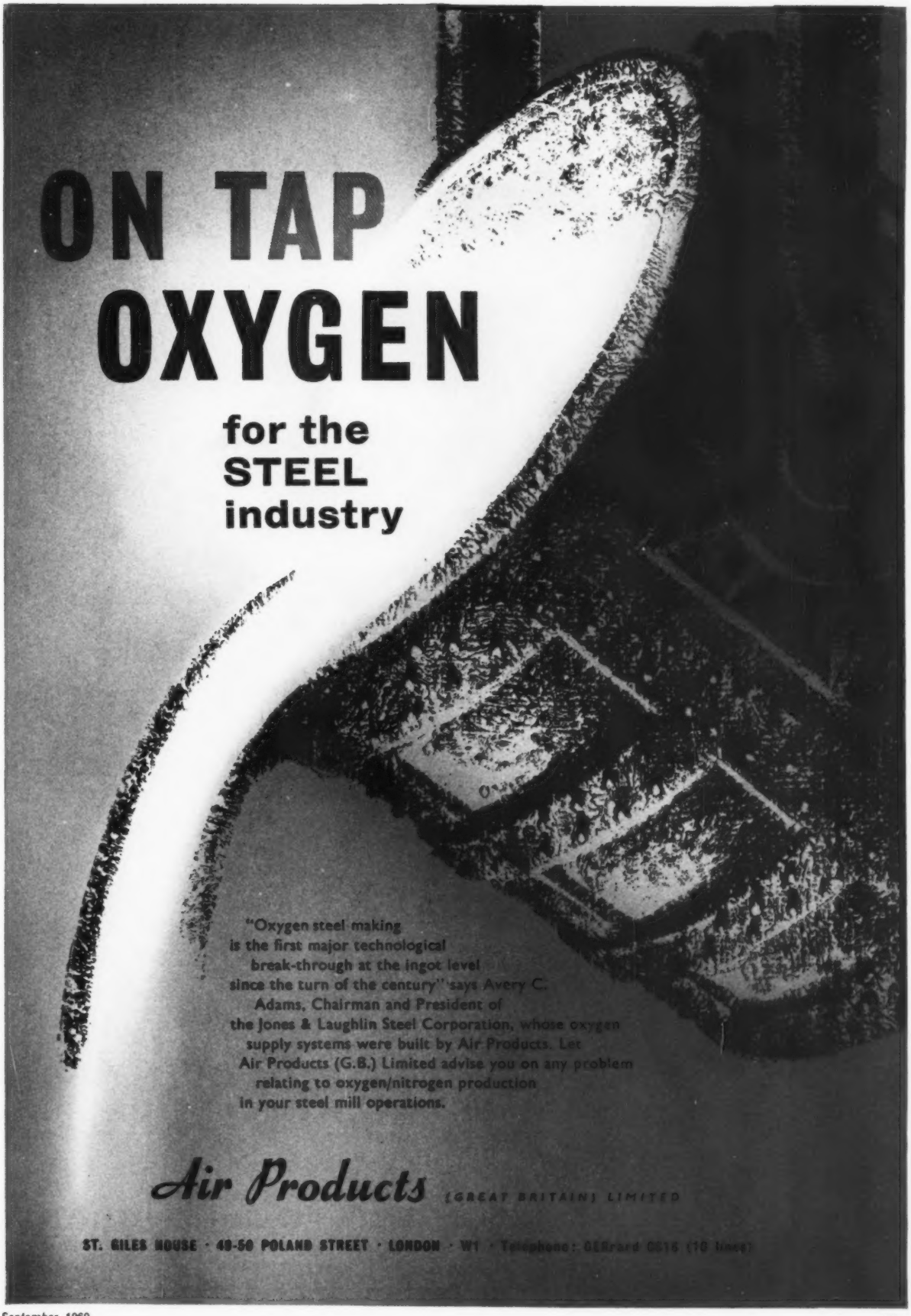
PARSONS are proud to be associated with the great new completely integrated Iron and Steel Works which RICHARD THOMAS & BALDWINS LTD., are installing at Llanwern near Newport. At these new Spencer Works, PARSONS (through PAROLLE ELECTRICAL PLANT CO. LTD. the main contractors) are to install:—

Three Turbo Blowers each delivering 170,000 cu. ft. of free air per minute at 45 psig.
Two 10 MW Turbo-Generators passing off steam at 175 psig. The steam conditions are (as on the Turbo-Blowers) 625 psig at stop valve with temperature 830 deg. F.

PARSONS turbo-blowers & turbo-generators

C. A. PARSONS & COMPANY LIMITED

HEATON WORKS · NEWCASTLE UPON TYNE · 6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for the
STEEL
industry
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is the first major technological
break-through at the ingot level
since the turn of the century" says Avery C.
Adams, Chairman and President of
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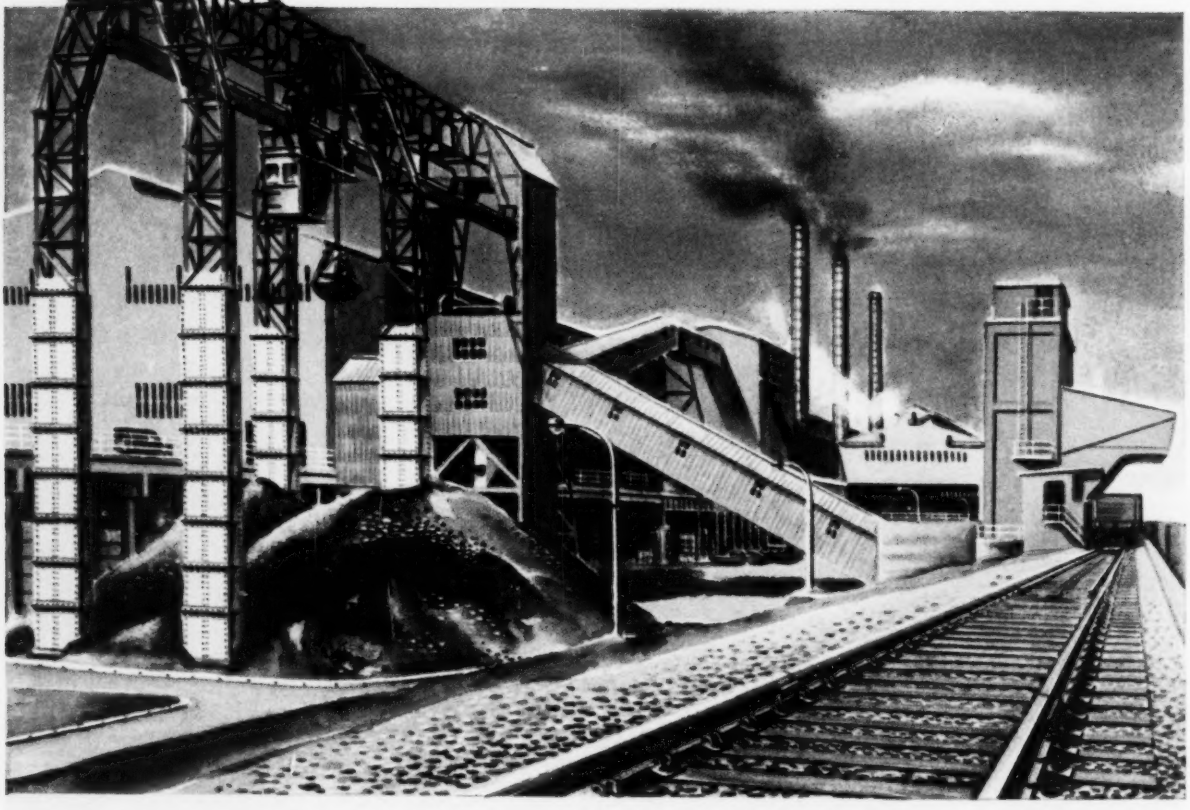

This raw material handling plant at Ravenscraig includes wagon marshalling, a dust trapping tippler, conveyors and a telpher, together with measuring box feeders and pan wagon control.

BY

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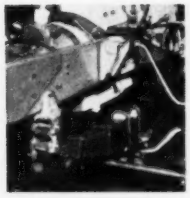

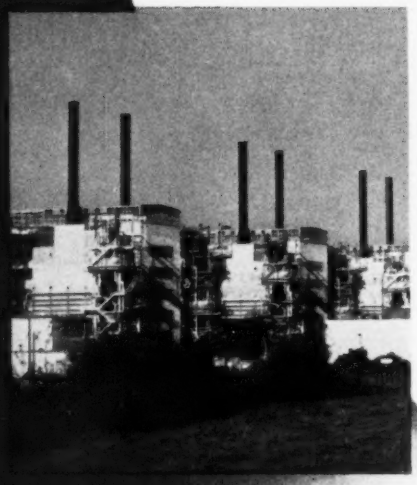

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26" VERTICAL MILL STAND

for a Continuous Billet Mill
Supplied together with pendulum shear
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SOMETHING NEW IN SHEET STEEL . . . ready surfaced with durable vinyl in the colour of your choice, all set for fabrication—that's **Stelvetite.** It can be cut, deep drawn, folded flat and welded; and the hard-wearing vinyl face which **Stelvetite** shows the world will resist oils, detergents, acids—and time. It comes in a wide range of colours and textures. It is made into vehicles, partitions, ceilings, furniture, kitchen equipment, wall cladding and drums . . . all of them colourful and durable. In production costs alone **Stelvetite** is a money saver. Investigate it!

Stelvetite

Plastic bonded to Steel

**Accepted by the Council of Industrial Design for "Design Index."*

All enquiries about Prestcold refrigeration units should be made to: Prestcold Division, Pressed Steel Company Ltd., Swansea, Glam.

STELVETITE*—made in co-operation with BX Plastics Ltd—**by John Summers & Sons Ltd.** who produce over 52,000 miles of sheet steel every year. Write to us at Dept. JS, Shotton, Chester, for full information.

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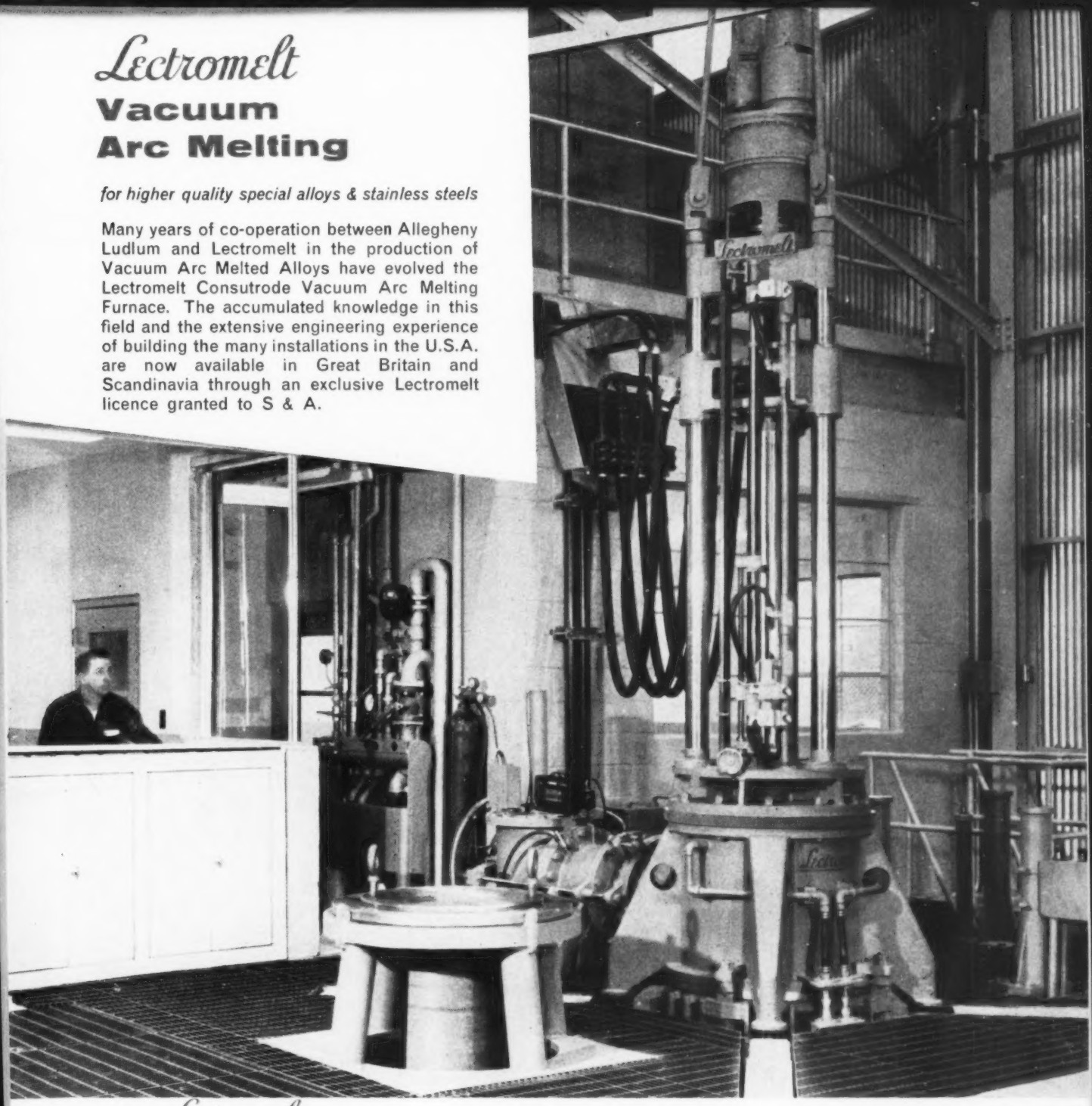
Vacuum Arc Melting

for higher quality special alloys & stainless steels

Many years of co-operation between Allegheny Ludlum and Lectromelt in the production of Vacuum Arc Melted Alloys have evolved the Lectromelt Consutrode Vacuum Arc Melting Furnace. The accumulated knowledge in this field and the extensive engineering experience of building the many installations in the U.S.A. are now available in Great Britain and Scandinavia through an exclusive Lectromelt licence granted to S & A.



A 20" x 8500 pound furnace installed by Lectromelt at Latrobe Steel Co., Latrobe, Pennsylvania.

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BRITISH STEEL
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RIMMING
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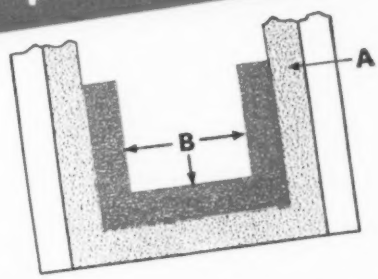
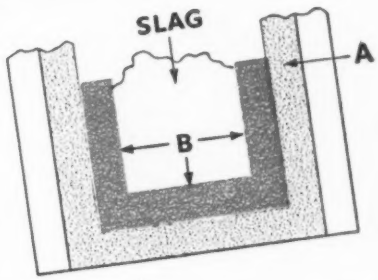
GR '341' DOLOMITE BRICK introduces NEW *Simplified* TECHNIQUE



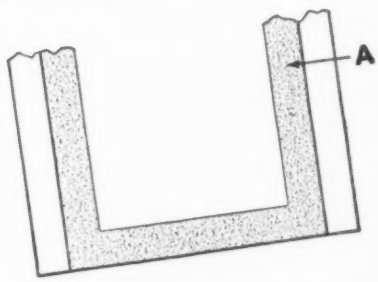
1 Existing Basic wall (A)
Protective GR '341' wall
and Hearth. (B)—Lay dry



2 Spray GR '341' wall with water to hydrate Dolomite Bricks (B)

NOTE *GR '341' bricks will powder to leave slag deposit free from main wall and hearth (A)*

3 Remove slag, and basic walls (A) will remain intact.



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Everything in Furnace Linings

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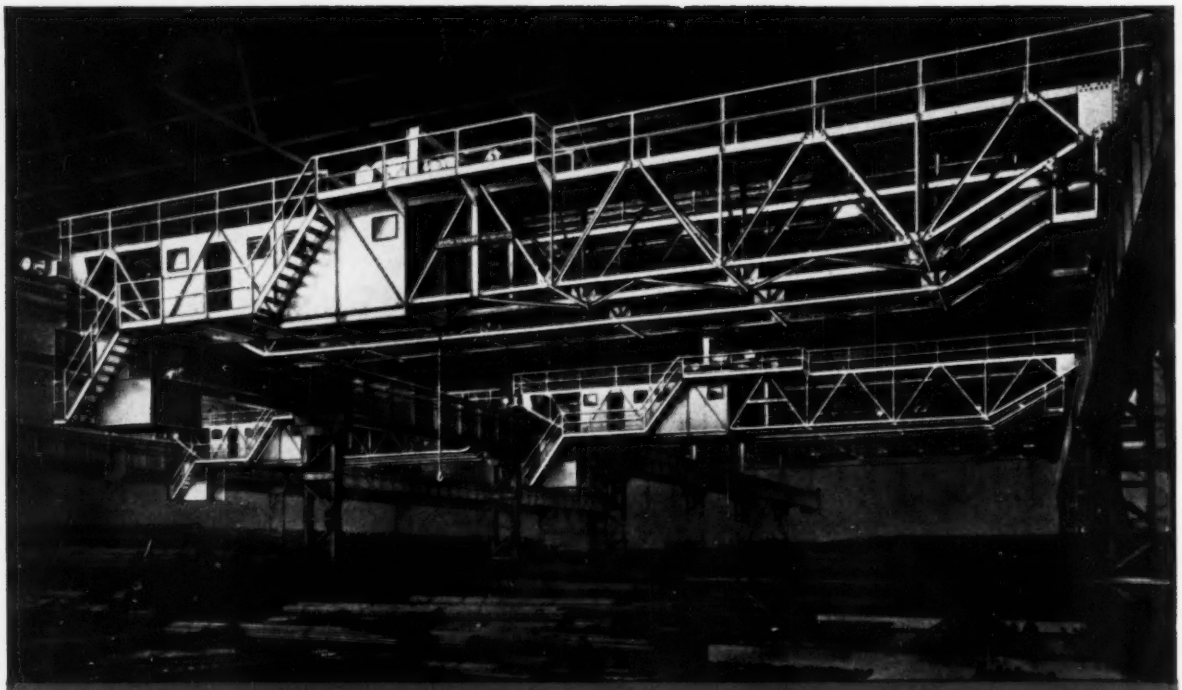
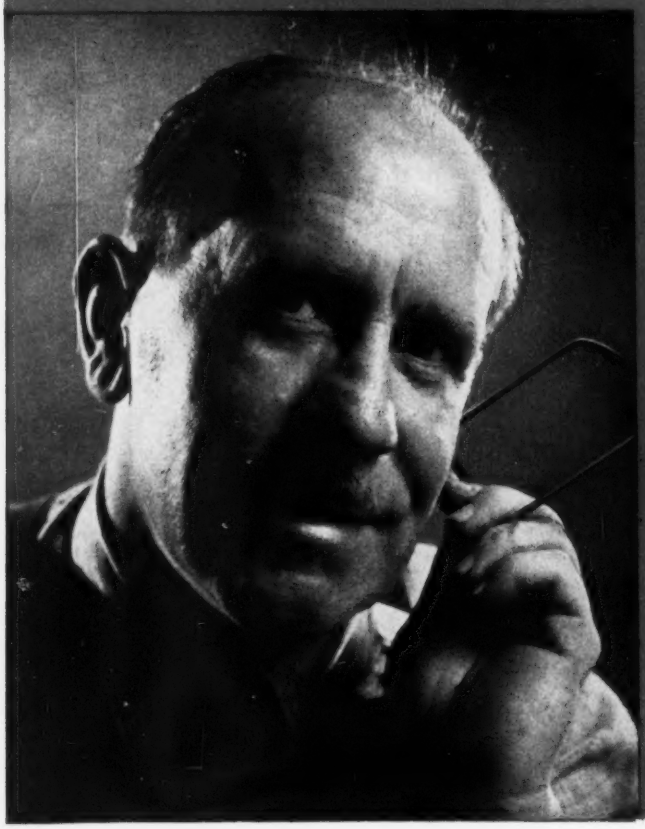


No problems **OVER** his **HEAD**..!

...he can rely on his...



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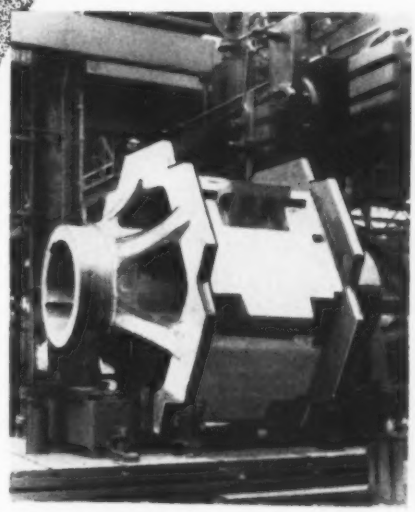
Best quality steel castings have many applications which offer certain economical advantages over other production methods. Before finalising your requirements **THINK OF CASTINGS FIRST** and consult Hadfields . . . . . a name for Steels in **CARBON—LOW ALLOY —STAINLESS—MANGANESE— WEAR RESISTING—HEAT RESISTING STEELS.**

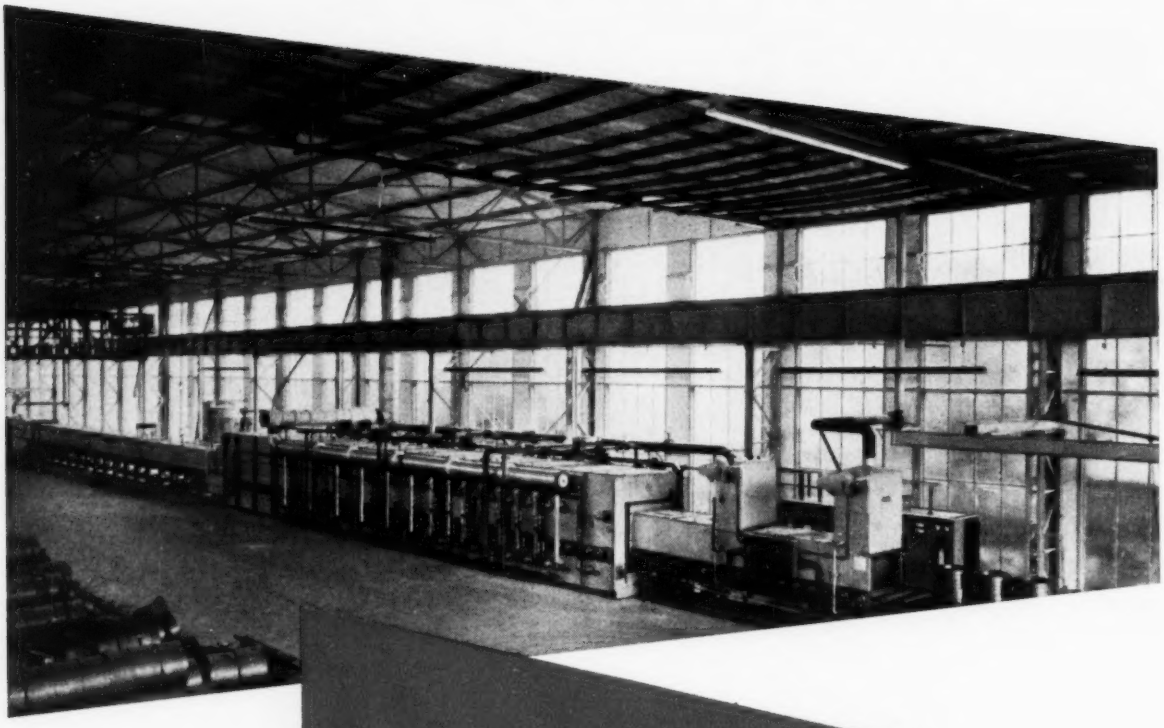


Top tumbler for tin dredging. Weight 17½ tons.

Tilting cylinder assembly. Weight 18 tons.

HADFIELDS
SHEFFIELD

HADFIELDS LTD., EAST HECLA WORKS SHEFFIELD ENGLAND



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INCREASE PRODUCTION AND QUALITY LOWER MAINTENANCE AND OPERATING COSTS, OUTPUT UP TO 5 TONS PER HOUR, GAS OR ELECTRICALLY HEATED

Suitable for:—

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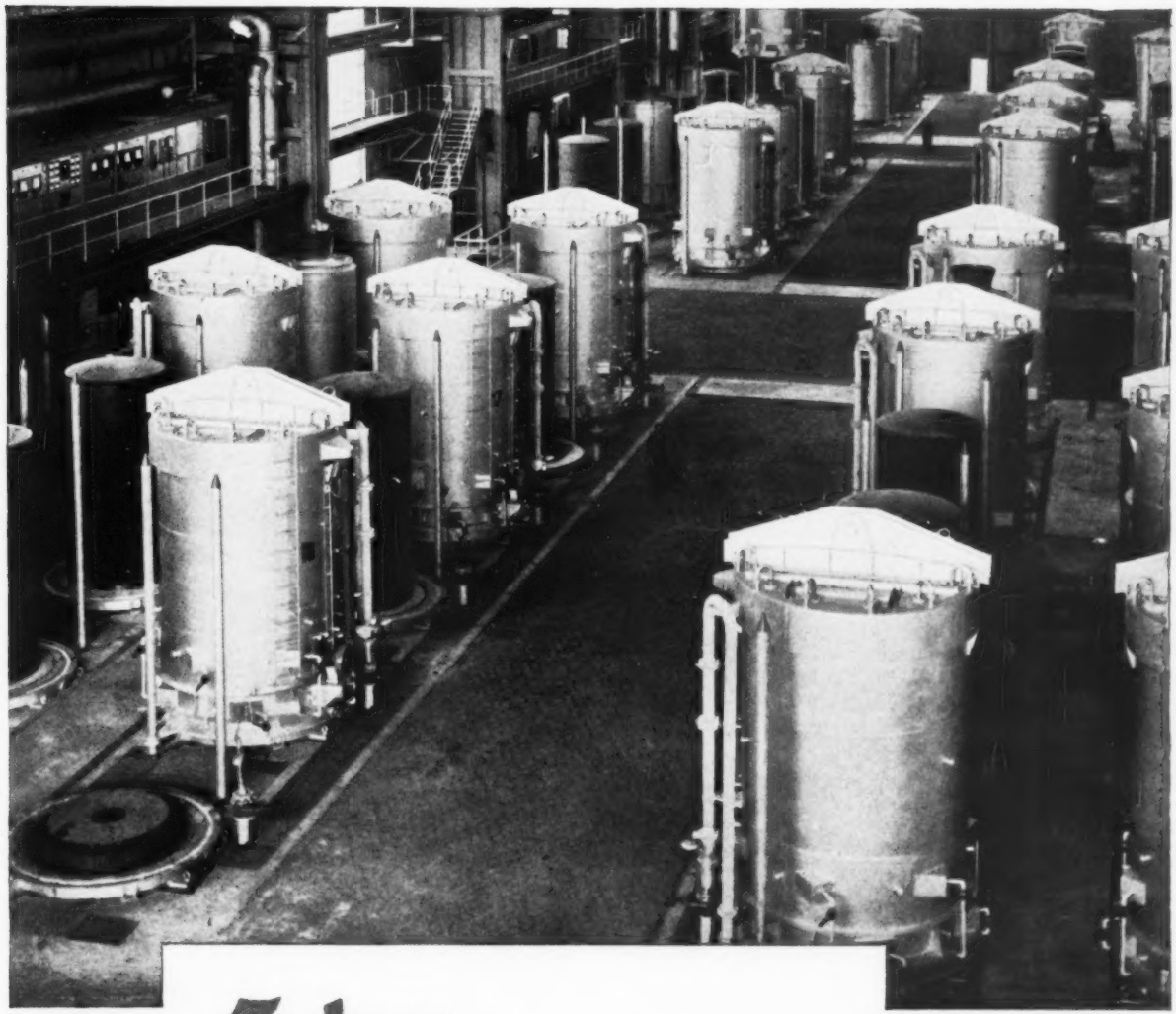

Salem for many years builders of modern multi-stack coil annealing plants now offer in collaboration with Messrs. Matthias Ludwig of Essen, the most up-to-date single stack coil annealing plants for both sheet and tin plate. 180 furnaces installed during the last eight years.

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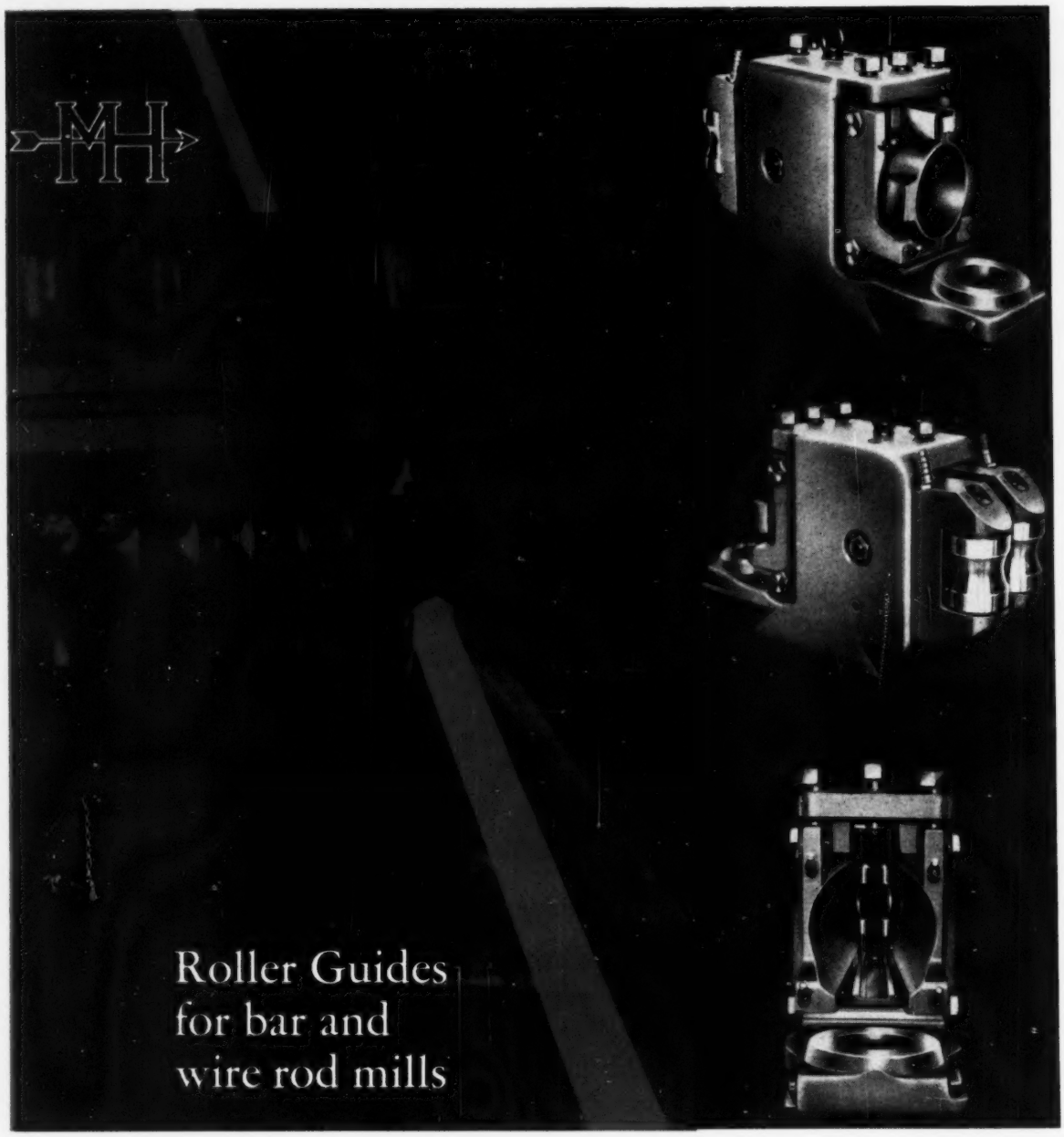

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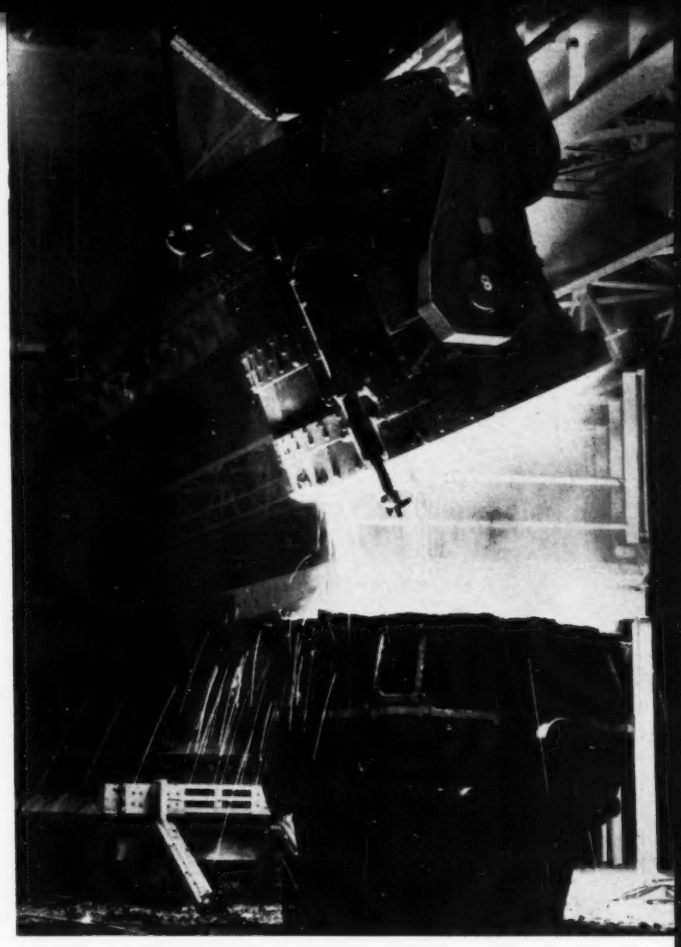

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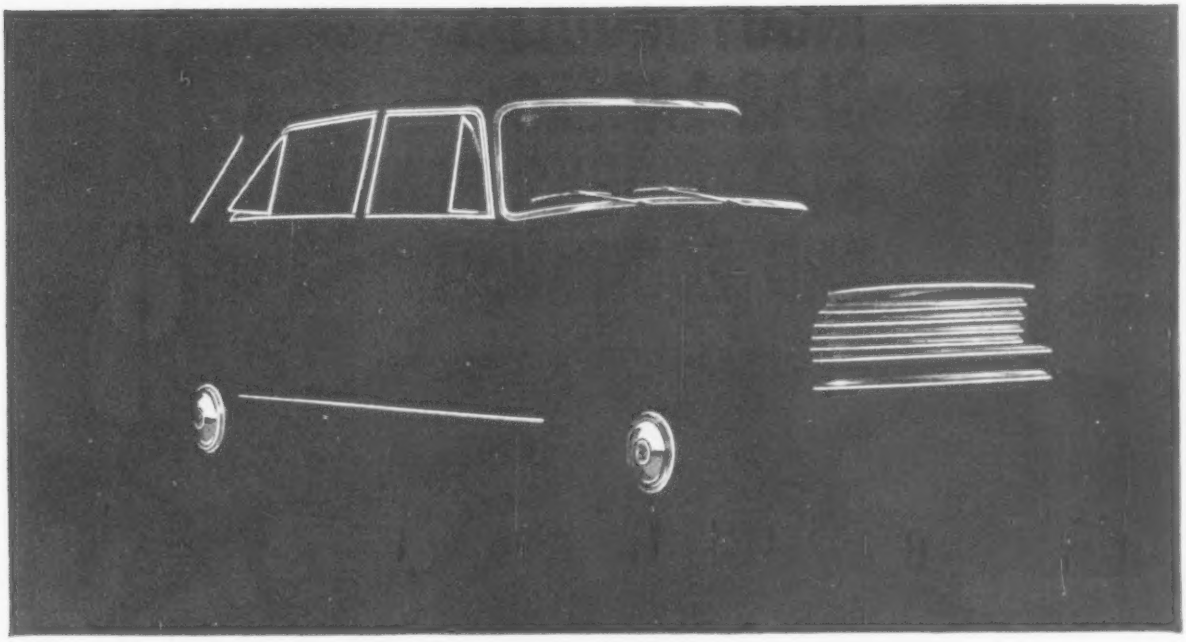

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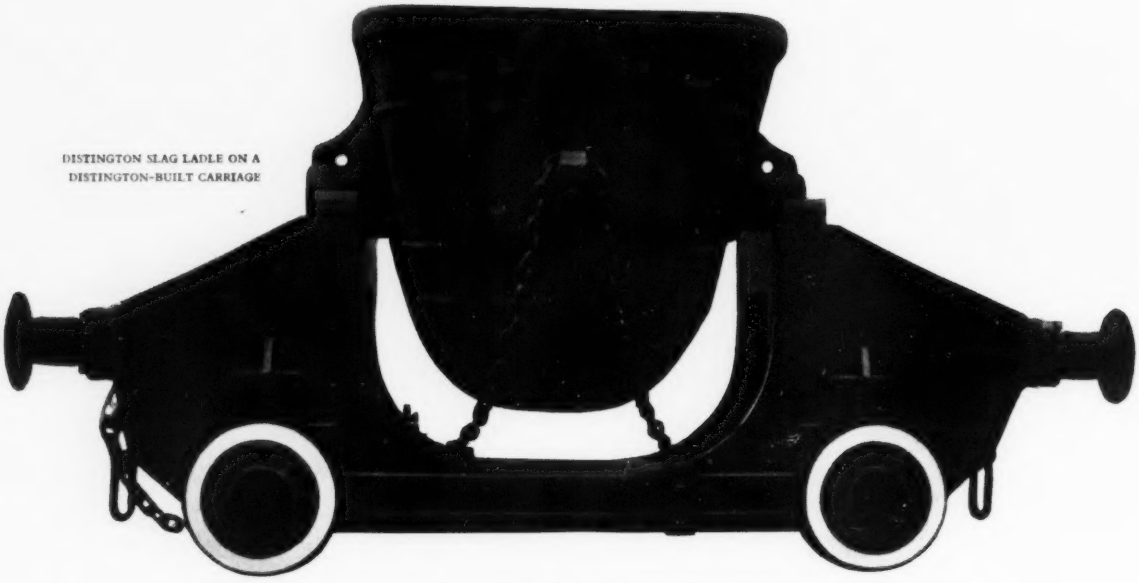
DISTINGTON SLAG LADLE ON A DISTINGTON-BUILT CARRIAGE



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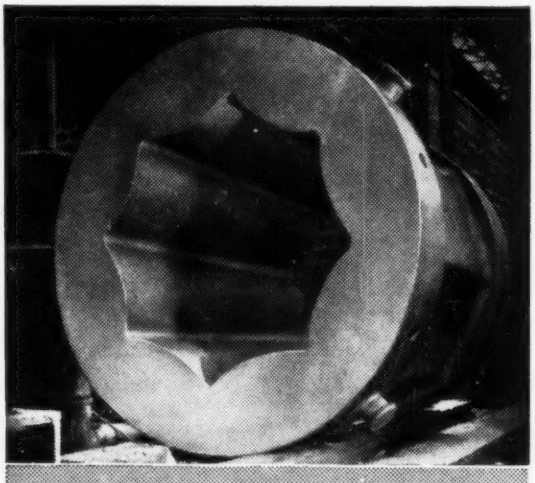

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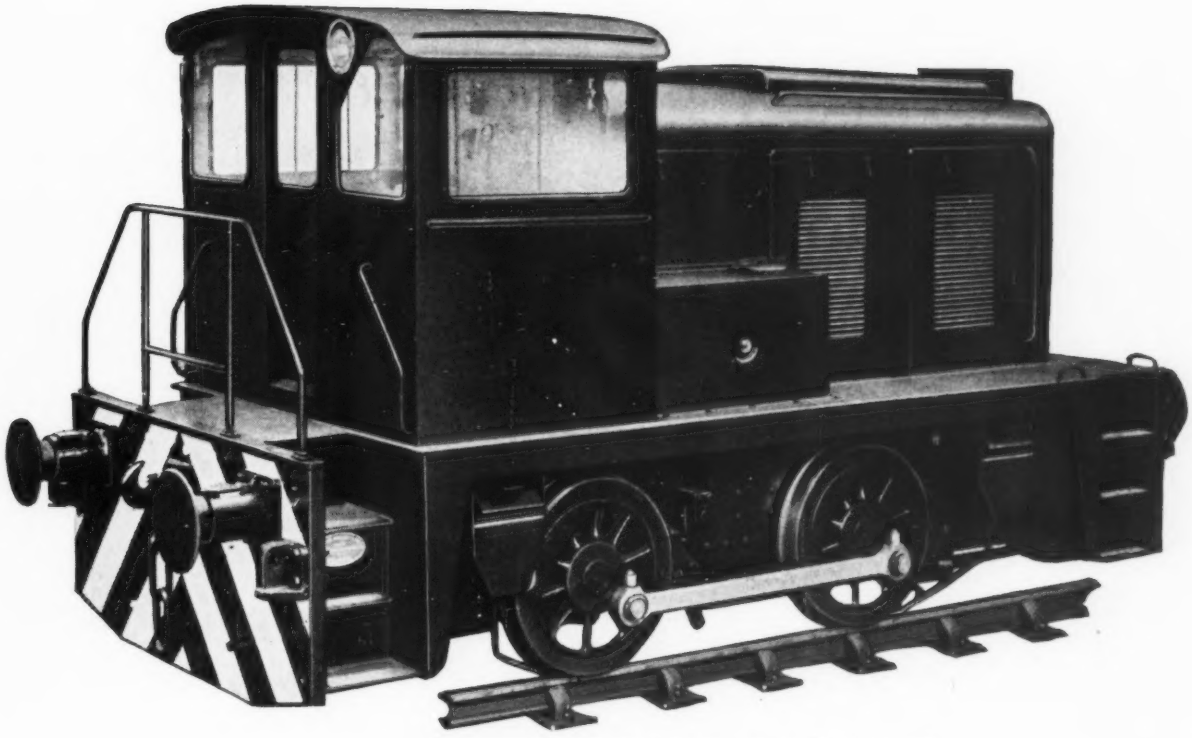

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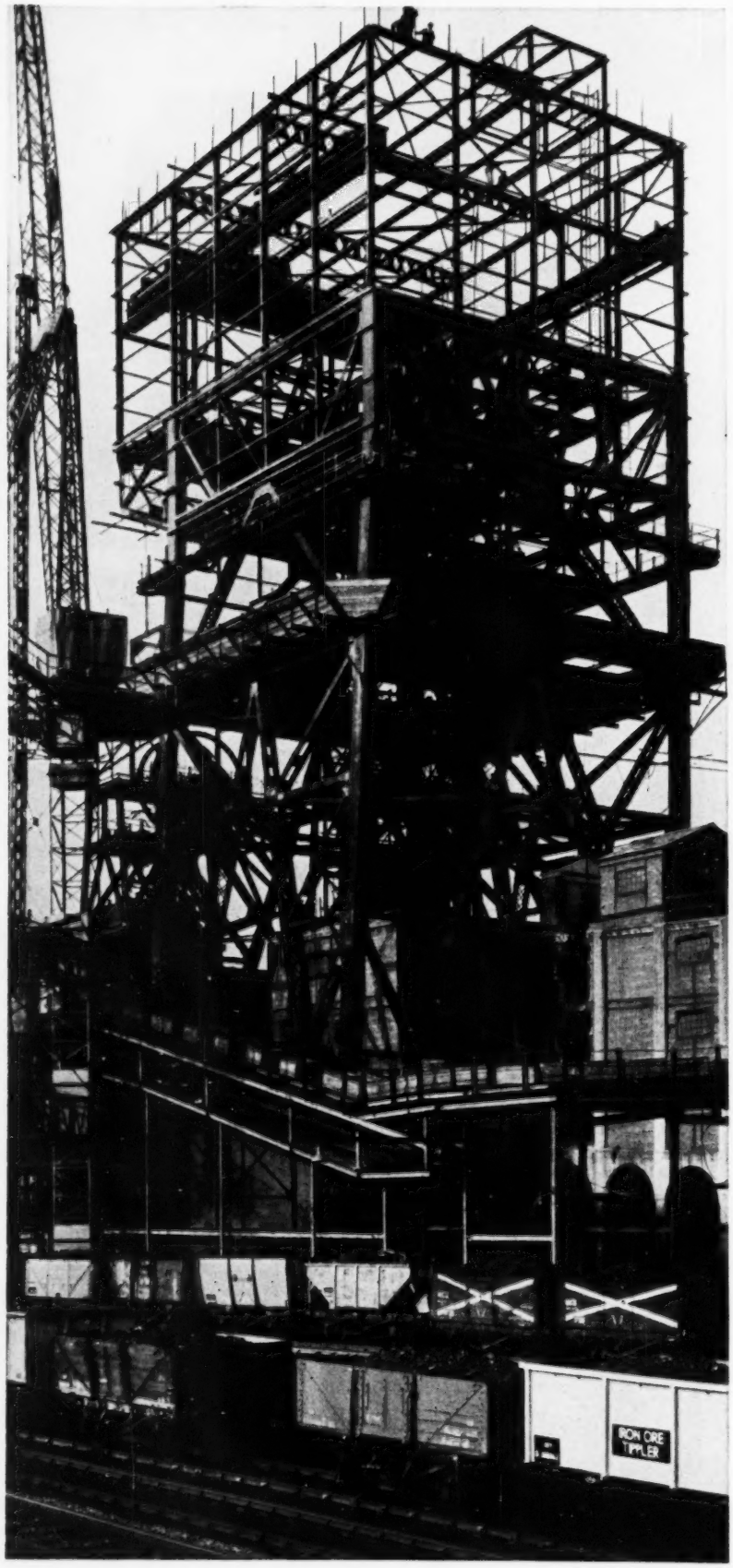

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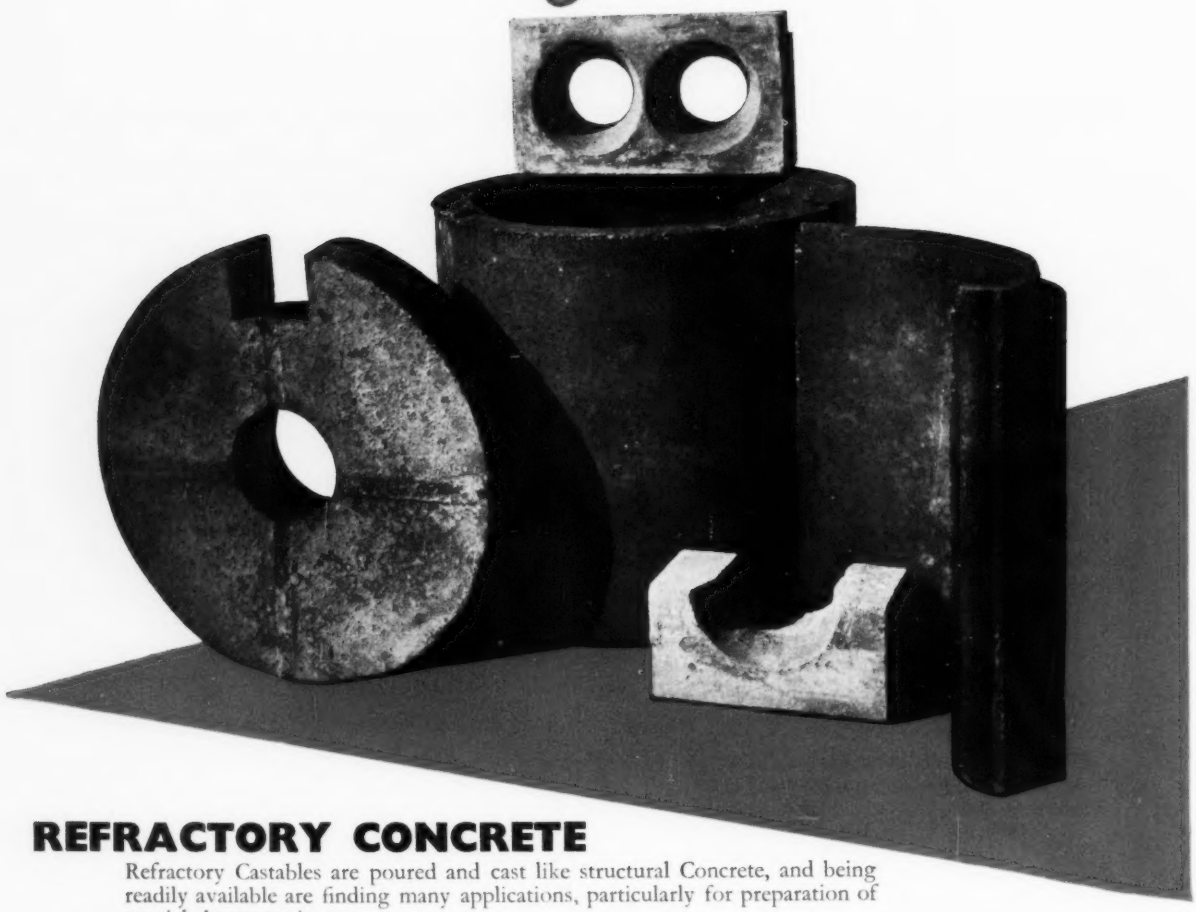

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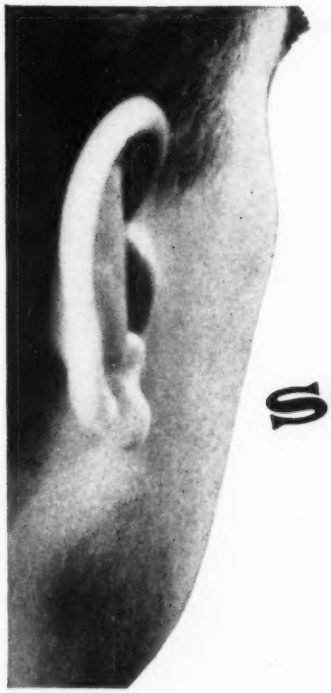

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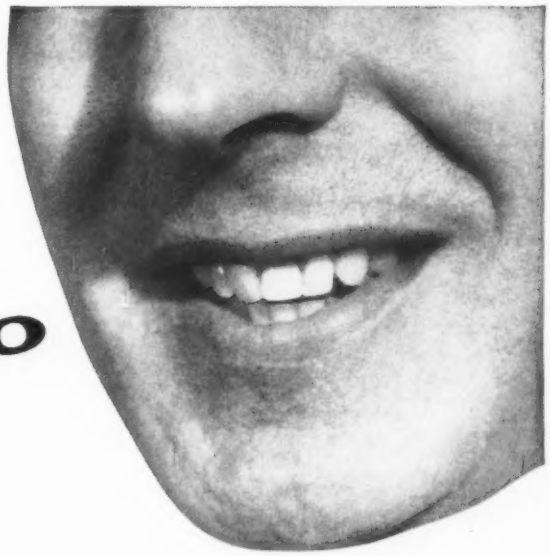

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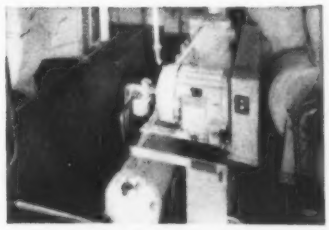

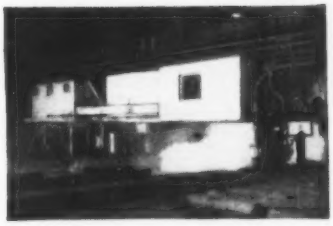

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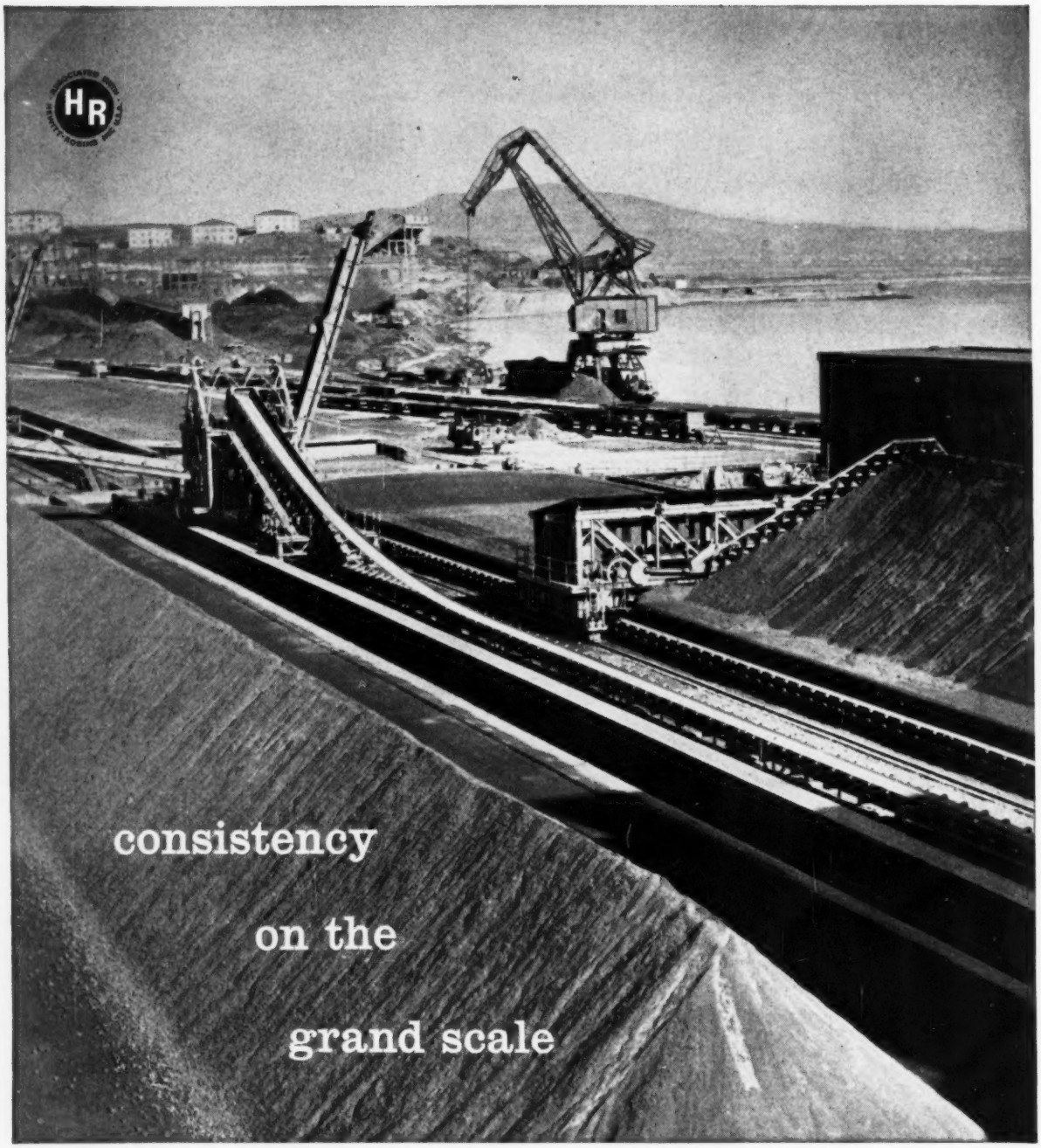

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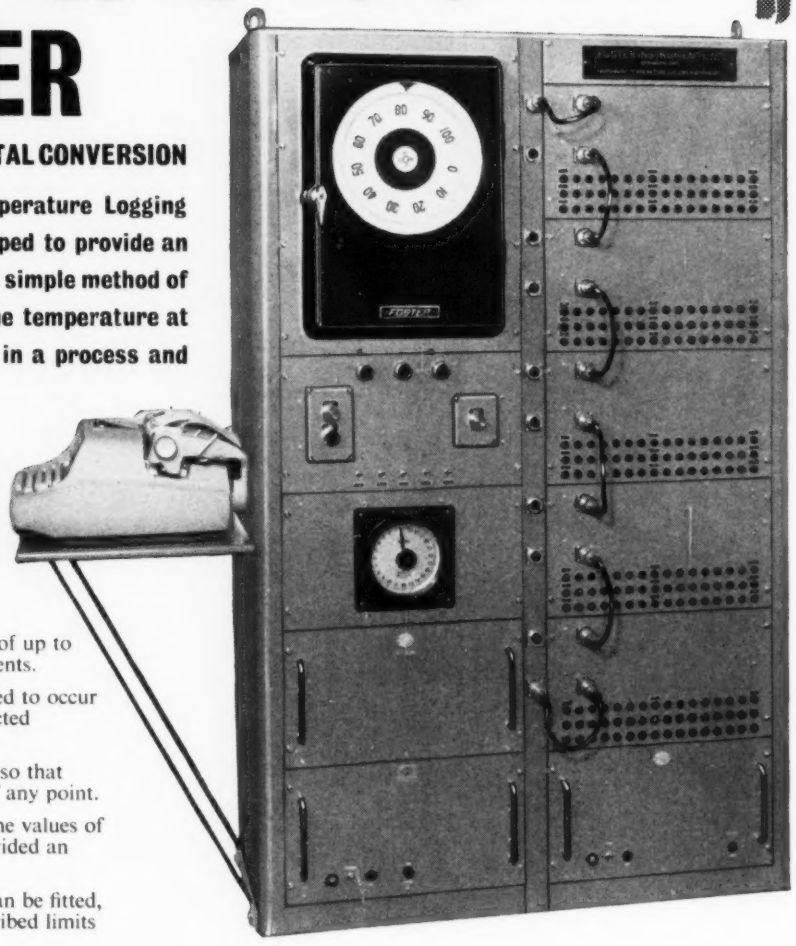

SOME OF THE CAPABILITIES OF THIS EQUIPMENT

Indicates and/or logs the temperatures of up to 208 thermocouples or Resistance Elements.

The scanning of stations can be arranged to occur at selected time intervals and for a selected number of the measuring points.

An external recorder can be connected so that a continuous record can be obtained of any point.

The equipment can be used to record the values of quantities other than temperature, provided an appropriate sensing element is used.

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The typewriter can be replaced by:

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3. a column numerical printing machine

APPLICATION

The equipment can be used with advantage in any process which the measurement at a large number of stations has to be recorded manually.

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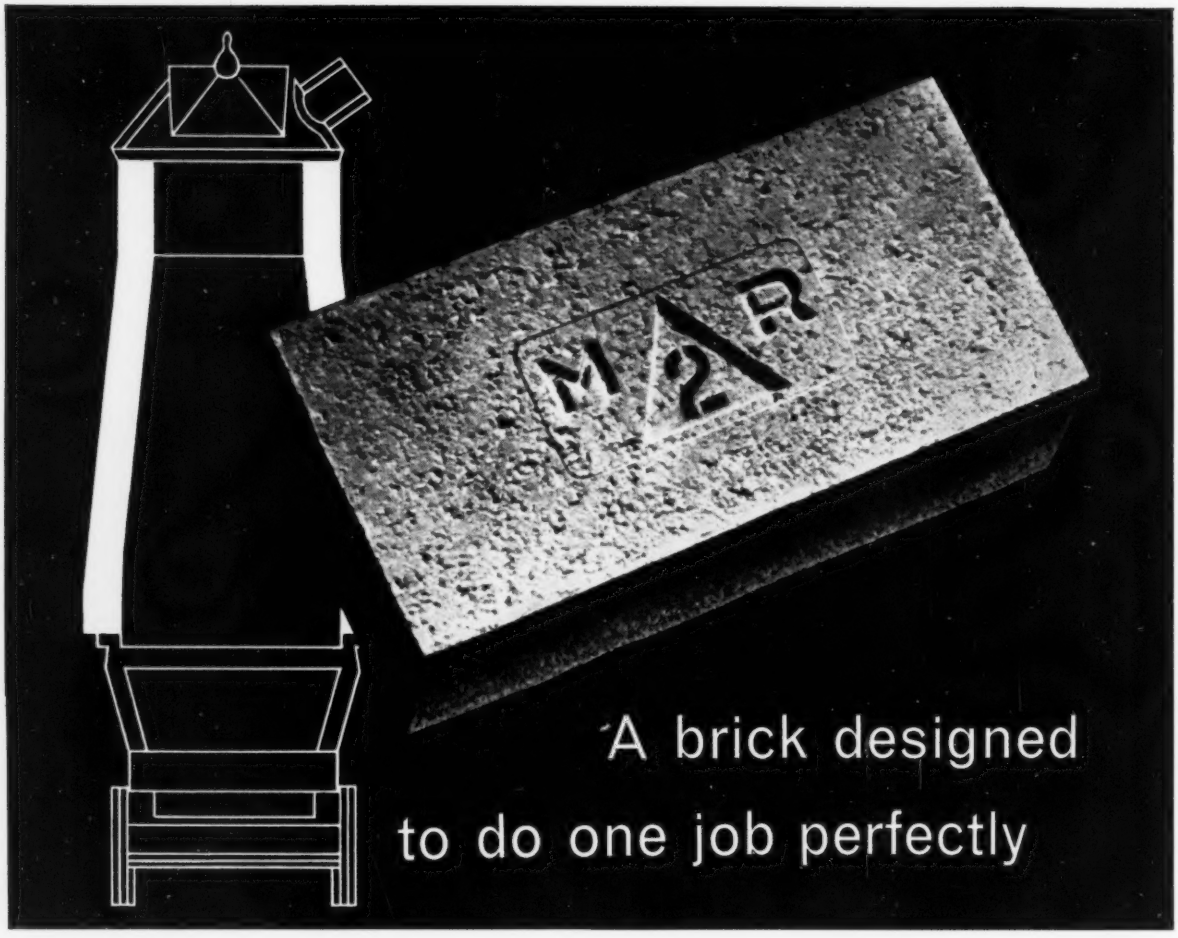

This is the M.R.2 Blast Furnace refractory – a very special brick indeed. It was developed for one application – the lining of Blast Furnaces operating on lean ores – with 100% sinter – and driven hard. M.R.2 is an extremely hard, low porosity refractory to enable it to resist the abrasive effect of the sinter. It is completely unaffected by carbon monoxide and is manufactured from very carefully selected raw materials prepared and moulded under the most stringent control with a final kilning at temperatures of the order of 1600°C to ensure complete volume stability at normal Blast Furnace operating temperatures. This is a brick with a narrow market, but it is typical of the way the Chemists, Physicists and Engineers at Neston look upon the making of refractories – as a branch of precise chemical engineering. Amongst the other things they make with equal care are M R.1 and M.R.60 bricks and special shapes; low-heat storage insulating refractories M.I.28, and the whole range of Tri-Mor Mouldables and Castables for Monolithic linings which are having such a profound effect on all classes of furnace design.

MORGAN Refractories Ltd

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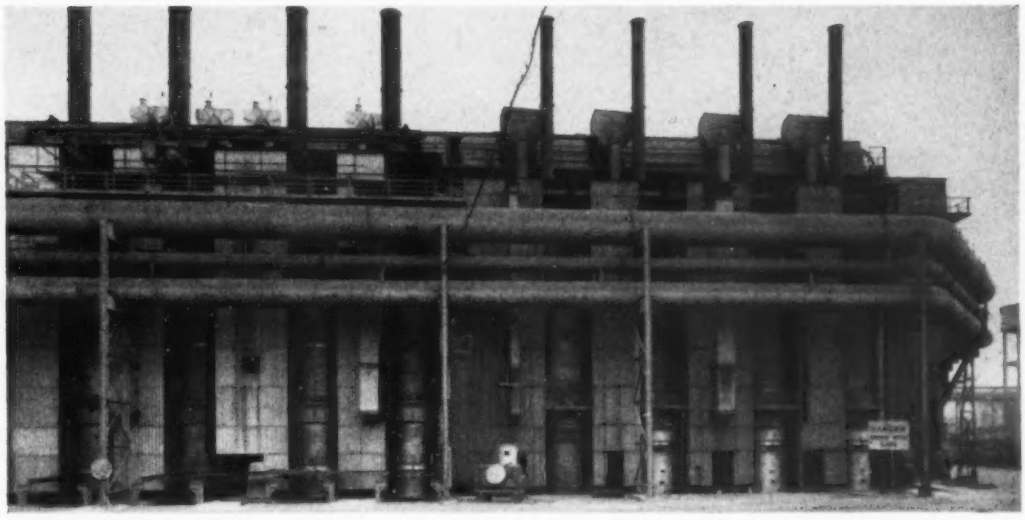
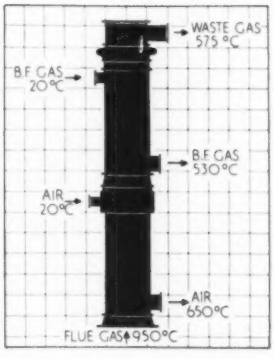
M.E./SCHACK RECUPERATORS FOR SOAKING PITS





The eight sets of M.E./Schack convection flue tube recuperators shown in the above photograph are mounted on eight one way fired soaking pits at the Appleby-Frodingham Steel Co., Scunthorpe. The recuperators are mounted in "tower" formation, the gas recuperator being directly over the air recuperator as shown in the sketch on the left. The latest four sets are designed for heating 170,000 c.f.h. of combustion air to 650°C, and 230,000 c.f.h. of blast furnace gas to 530°C.

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The next time you are working out your fuel problems, remember Bowaters. For their pulp and paper mills in Kent and Cheshire they chose coal—one of the factors that point to a buoyant future at Bowaters. The same decision made by you may well begin a success story that could make headline news—on Bowaters' newsprint.

PROGRESSIVE INDUSTRY IS GOING FORWARD ON

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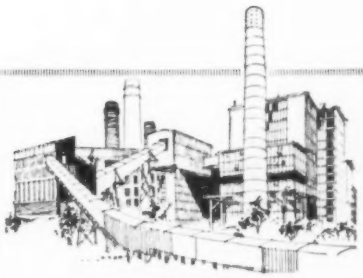

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Mr. L. A. Robey, Deputy Power Engineer, comments:

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The new pulverised fuel boiler which is designed for a continuous evaporation of 450,000 lbs/hr. at a pressure of 1,550 p.s.i.g. and a final steam temperature of 925°F. has steamed on coal from its commissioning date. The test efficiency obtained at full load was over 88% and this has been maintained consistently over long periods of operation with very high availability. The boiler stays very clean and with nominal maintenance excellent combustion conditions are maintained. The remote control is very flexible and meets the changing process demand without difficulty. Grit arresters collect the dust and discharge it automatically for disposal with the minimum amount of trouble.

The quality of fuels received has always been consistent and we have never failed to maintain the large stock piles which are required for the plant.

Bowaters' chose coal many years ago for their main fuel requirements at Kemsley—and still do."

Here are the details of the Bowater Kemsley boiler installation

Boilers: ..... 10 water-tube
Method of firing: ..... Chain grate stokers and pulverised fuel
Boiler efficiency: ..... 80% old boilers, 88% new boilers
Maximum steam rating: 940,000 lbs. per hour
Maximum steam pressure: 1,500 lbs. per sq. inch
Maximum steam temperature: ..... 915°F.
Annual fuel consumption: 320,000 tons of coal *(Kemsley alone)*

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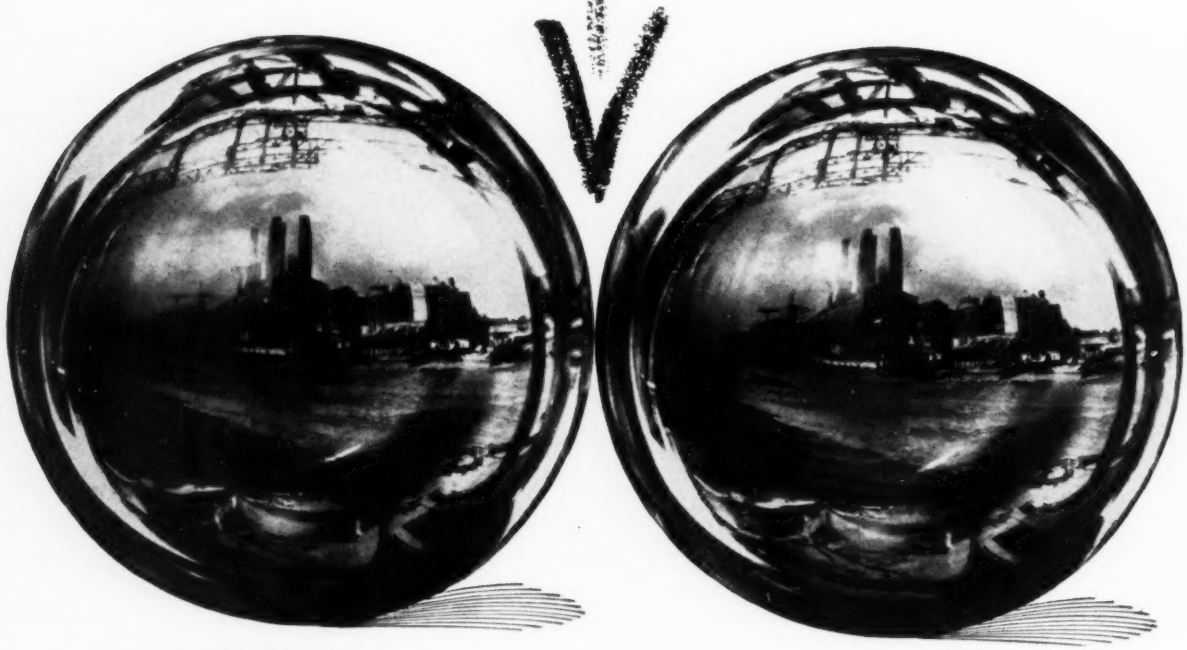
MOBILUX GREASE



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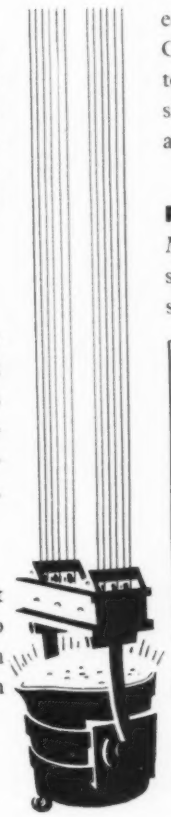

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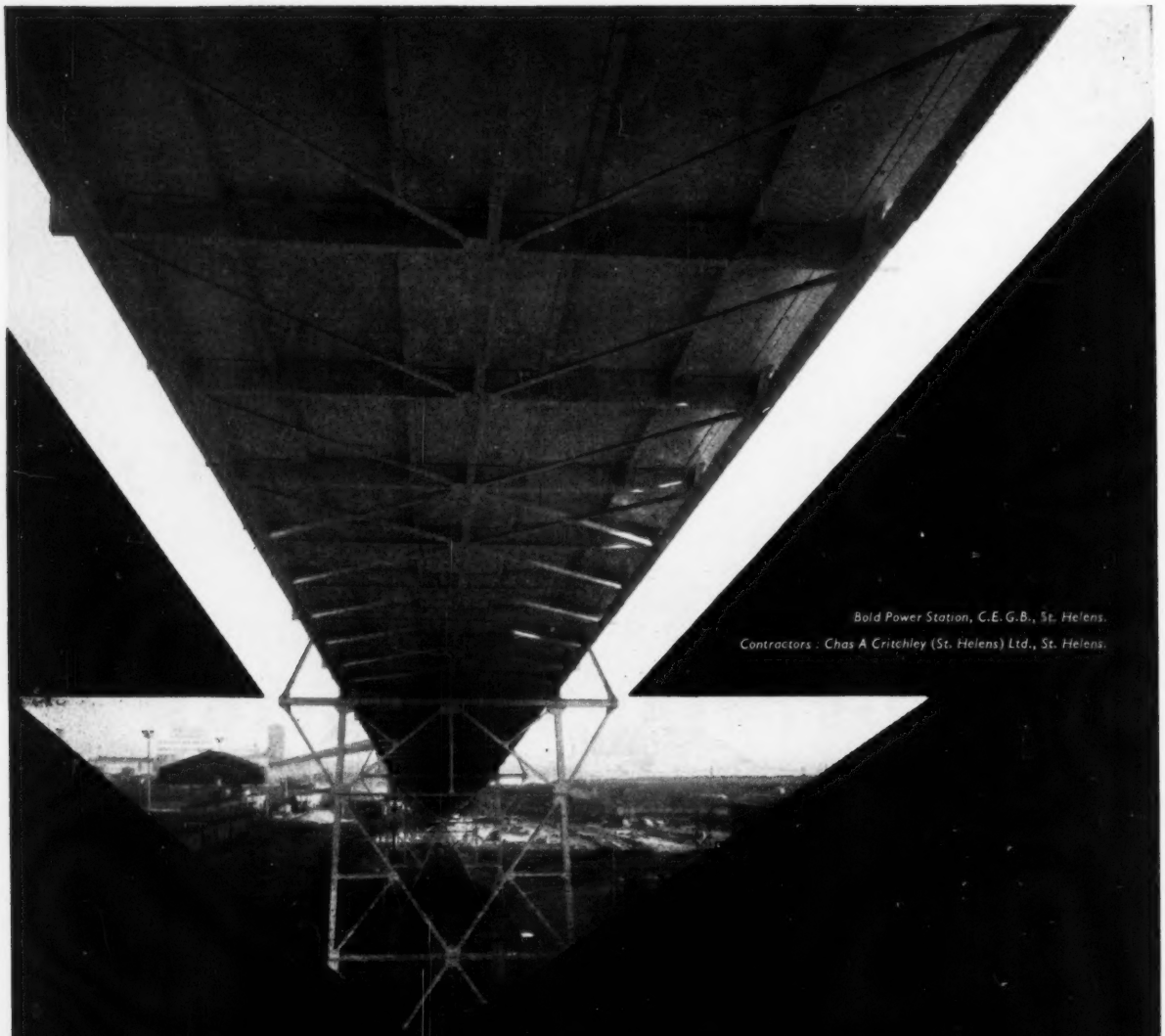

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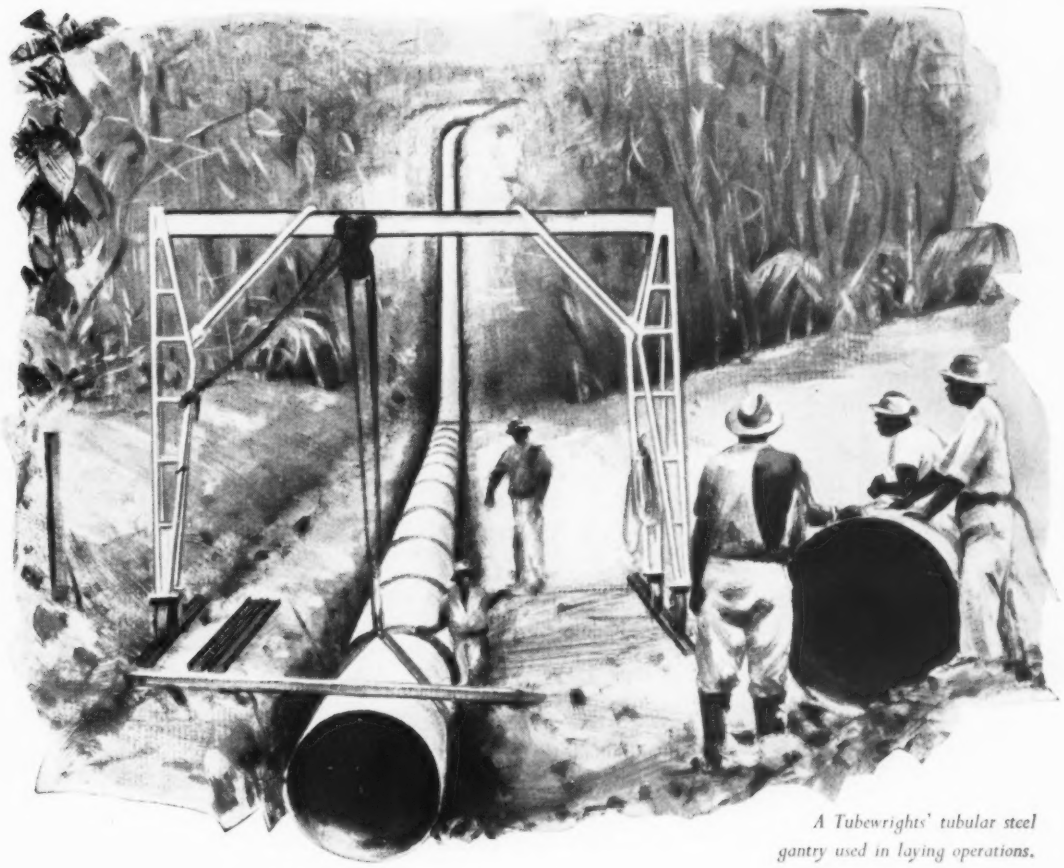

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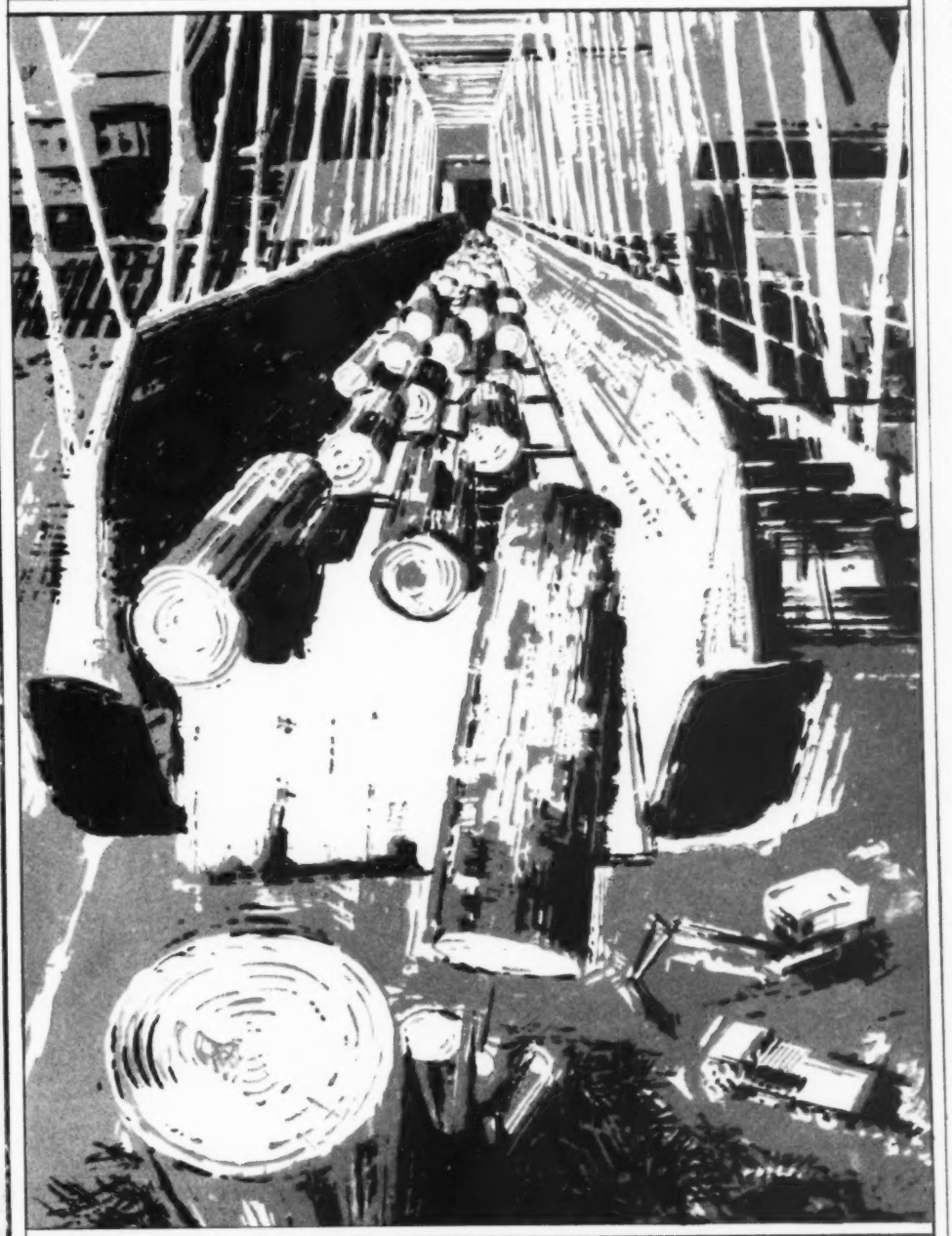

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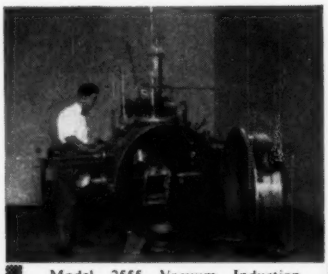

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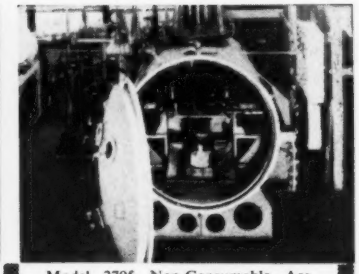

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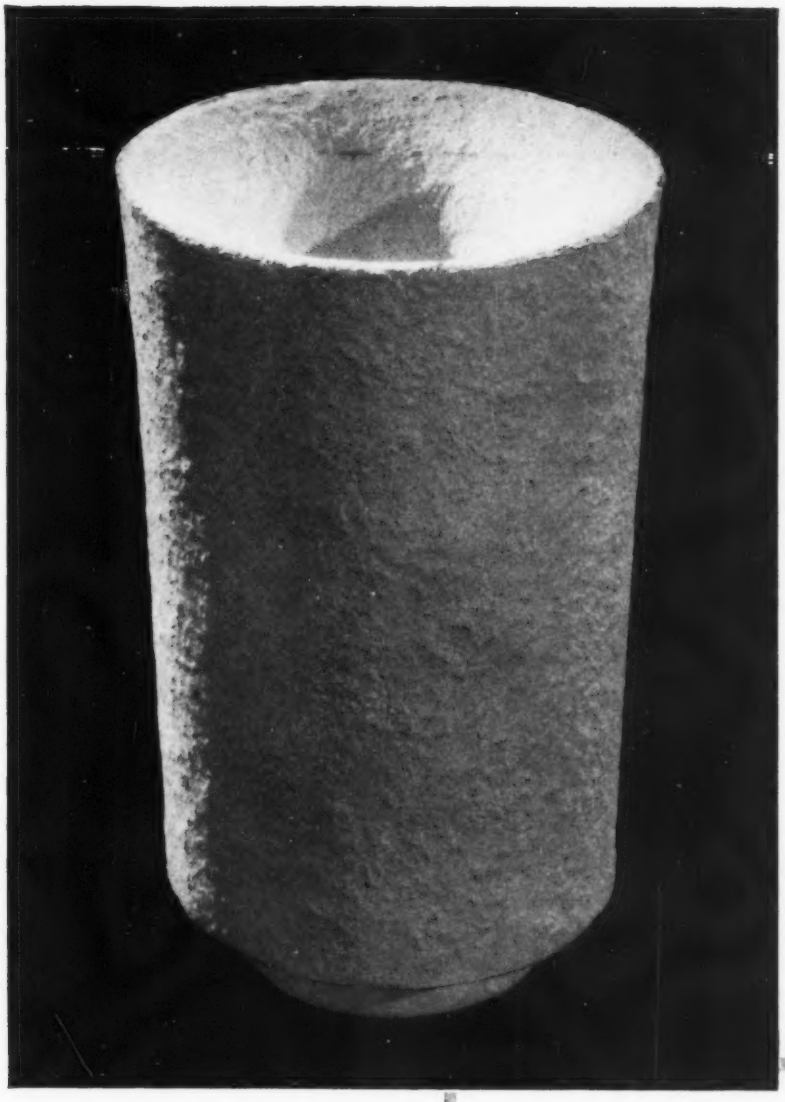

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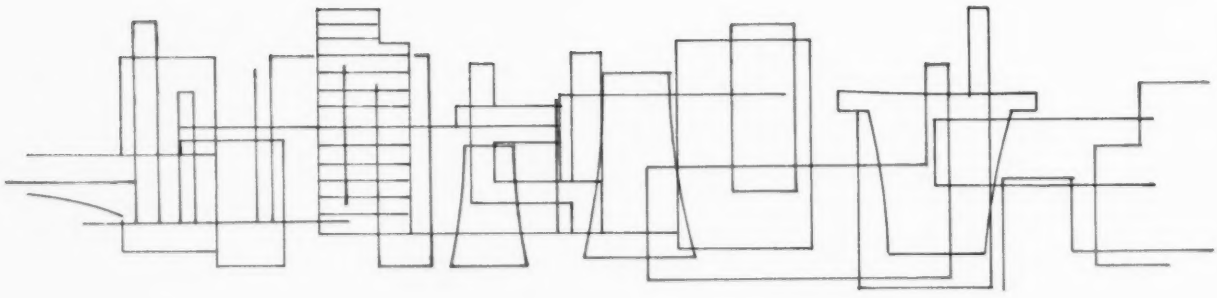

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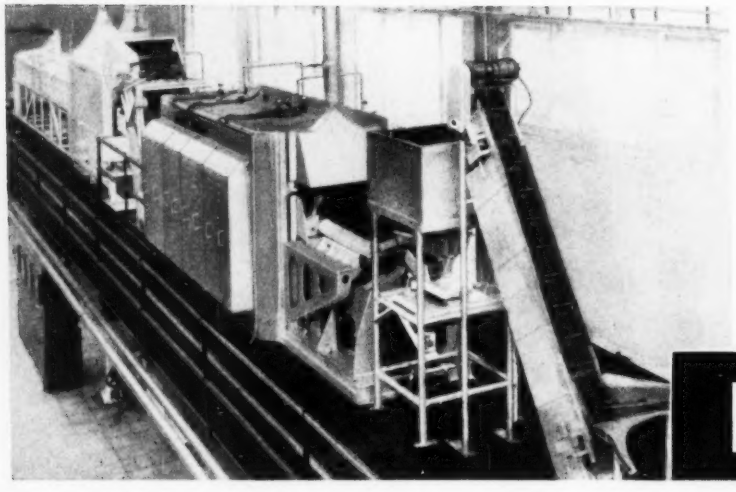

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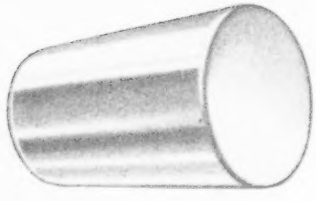

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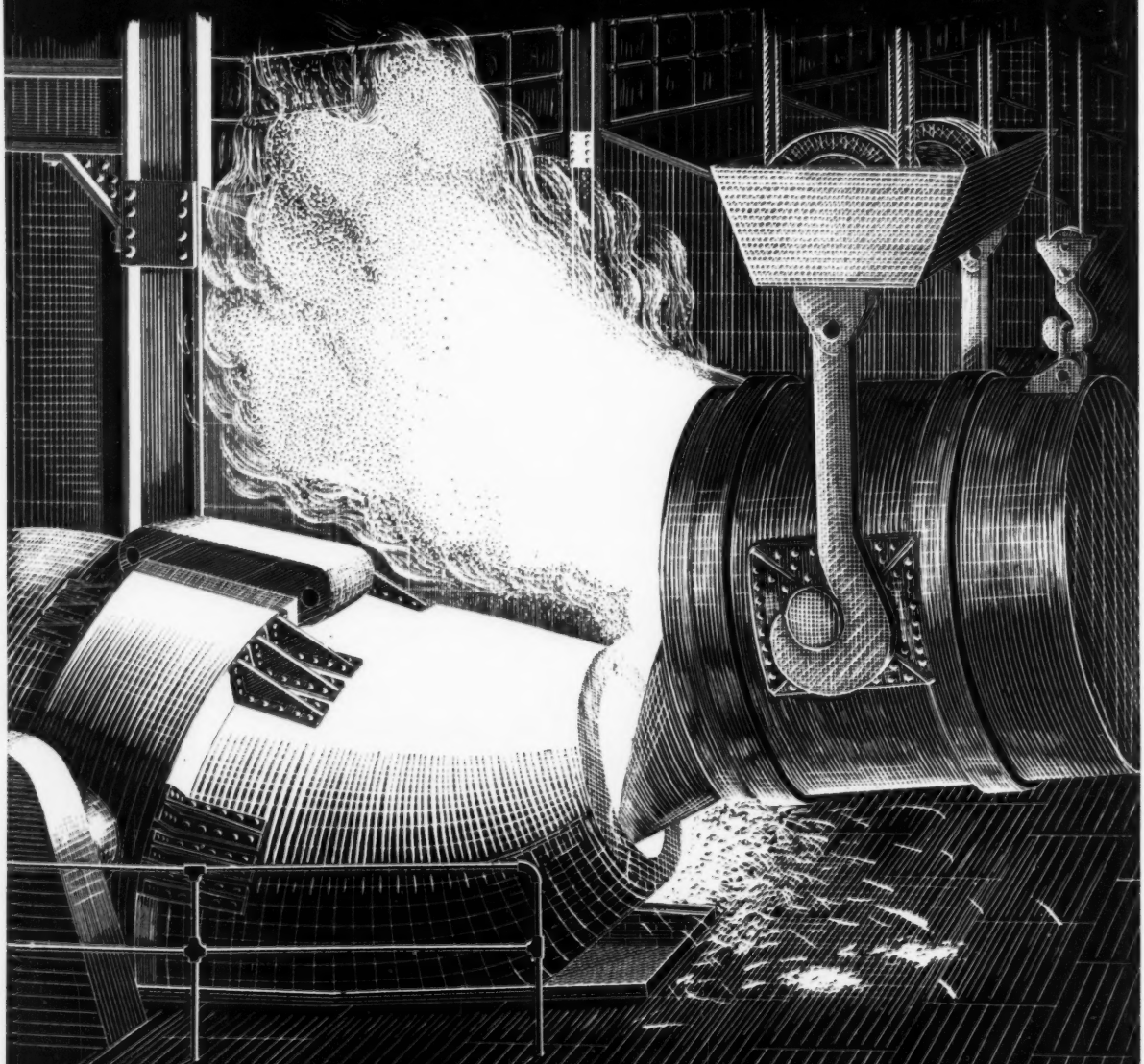

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MAJOR W. R. BROWN, D.S.O.

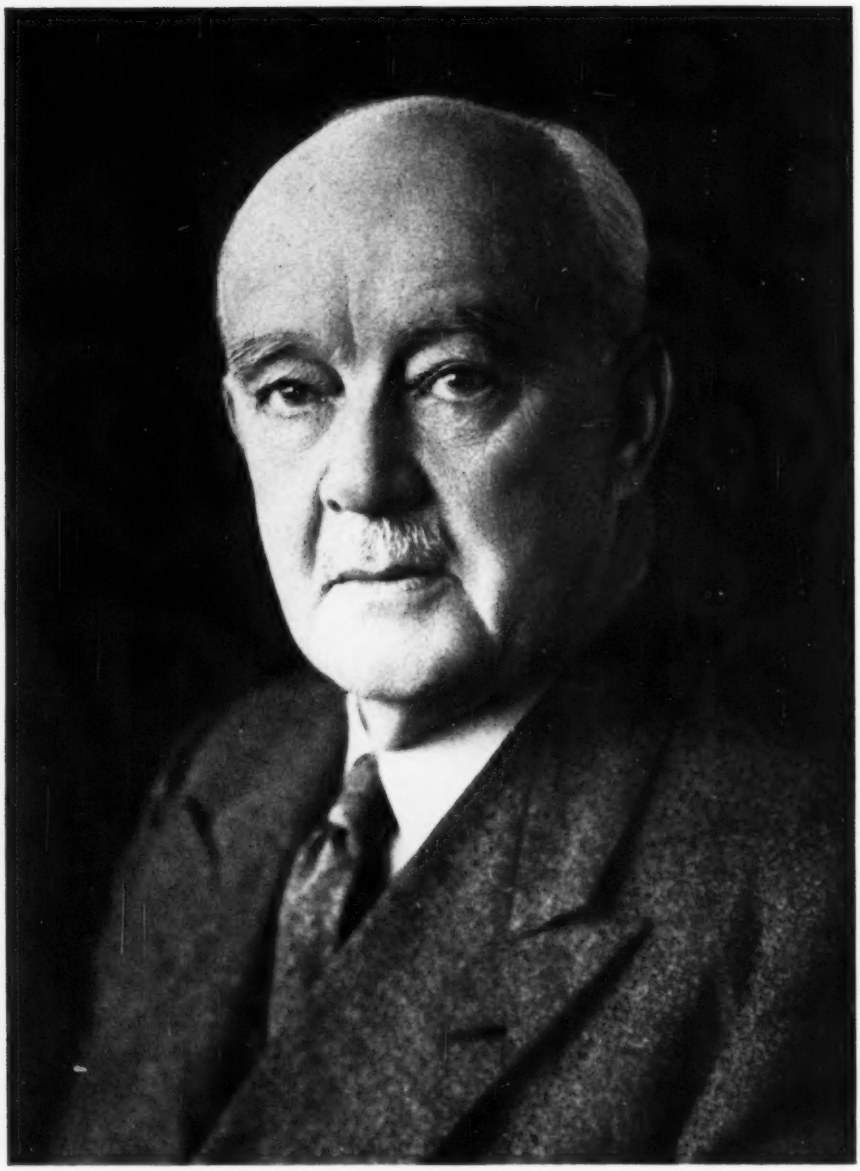
WILLIAM REID BROWN was born in 1888 and educated at Tonbridge School. In 1907, at the Cardiff Works of GKN he came under the guidance and tutelage of A. K. Reese, the leading blast-furnaceman of his day. Reese gave him a zest for furnace work which has lasted a lifetime. Major Brown served as an Artillery officer throughout the 1914–18 war; he was mentioned in despatches and awarded the D.S.O. After the war he was for a time with the Ministry of Munitions, then he returned to blast-furnace work, though this was frequently interrupted by a war injury.

In 1928 he joined Ashmore, Benson, Pease and Co. Ltd to modernize their blast-furnace building activities. A series of agreements with US designers led to the building of several notable furnaces and today some 50 stand to the credit of the organization under his control.

In 1942 he was made a director of Ashmores and two years later he was appointed to the board of the parent company, The Power-Gas Corporation Ltd, of which he subsequently became chairman and managing director. In 1960 this business amalgamated with Davy United Ltd, and Major Brown, retiring from active executive work, joined the board of the new parent company, Davy-Ashmore Ltd. He maintains his interest in blast-furnace matters by also serving on the board of the Ashmore, Benson Pease subsidiary.

He is a Past-President of the Cleveland Scientific and Technical Institution. He was on the Council, and is now an Honorary Life Member of the Cleveland Institution of Engineers. He was a member of the initial committee of the Engineers Group of The Iron and Steel Institute and has served on BISRA Plant Engineering Committees.

Major Brown joined the Institute in 1929. He was elected to the Council in 1959, and in 1960 he was made an honorary Vice-President.



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Aerodynamics of the open-hearth furnace as a means of controlling the processes of combustion and heat exchange

M. A. Glinkov

SYNOPSIS

This paper gives an account of Soviet research on the aerodynamics of an open-hearth furnace and their effect on combustion and heat-exchange processes. Extensive use has been made of models and the results obtained provide a scientific basis for OH port design. 1852

INTRODUCTION

V. E. Grum-Grzhimailo was the first person in Russia to use models to study the movement of gases in furnaces; he did this as early as the beginning of the present century although the theoretical fundamentals of using models of thermal devices based on the theory of similarity were developed later by Kirpichev and his school.[1]

In the USSR, models for studying the movement of gases in open-hearth furnaces with various types of port (Siemens, Venturi, Moll, and Maerz) were first used by Kuz'min in 1930–32.[2–7]

Later, the model method began to be used to study the movement of gases in furnaces of various types. A general theory of this work was published in 1936 in special monographs.[9, 10] On account of the special importance of so-called jet movement in the furnace chamber, many investigations were later carried out relating to the collision, and mixing of free and confined streams. The general theory of these investigations is presented in the third chapter of the book 'Fundamentals of the general theory of the thermal operation of furnaces'.[28] On the basis of research work carried out on jets, an attempt was made to describe combustion and the phenomenon of the bath-covering capacity of an OH flame, in mathematical terms.[11–13]

Since the war, the use of models for studying the movement of gases in furnaces has greatly increased in the USSR and the bibliography on this question now amounts to several dozen references. Results obtained from furnace research with models are widely used in the design and reconstruction of furnaces. More than this, when new types of furnace are designed, preliminary aerodynamic model studies are now practically compulsory. In the higher metallurgical educational institutions in the USSR, fundamentals of the theory of similarity and the use of models are being taught and suitable laboratories have been set up.

Research laboratories for model work exist in many research institutes and plant laboratories. Visual observations and photography are used in connexion with fluid (water) models.

Manuscript received 22 June 1960.

Professor Glinkov is Head of the Department of Metallurgical Furnaces at the Moscow Steel Institute.

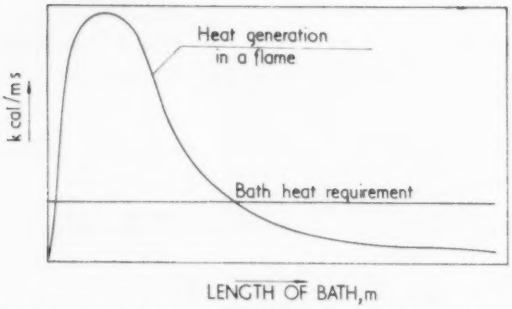
Velocities, pressures, and concentrations are measured by means of gas and air models. Frequently, these investigations are accompanied by parallel research with hot models or combustion rigs and actual furnaces. It has been shown that the hydrodynamic picture of an actual furnace can be established with adequate precision by trials on cold, liquid, and gaseous models.

With OH furnaces, special attention has been devoted to the influence of various port designs on flame aerodynamics. The part played by oxygen jets and compressed air introduced through the gas port and other aspects of melting-chamber aerodynamics have been studied.[14–27,36] An analysis follows of melting-chamber aerodynamics based on work carried out in the USSR mainly in cooperation with industry.

The aerodynamics of flames is a question of the greatest importance, since it governs:

(i) mass transfer processes which ensure heat generation
(ii) the removal of cooling gases from the bath surface and their replacement by fresh portions of a burning stream.
(iii) mass transfer processes associated with bath feed oxygen for technological purposes and the removal of the gases released from the bath surface.

In the long run, control of flame aerodynamics is a means of controlling both combustion and heat-exchange processes and, partly, the technology, so that it is essential first of all to formulate aerodynamic requirements in the melting chamber, starting from rational conditions for the development of combustion and heat exchange.



1 *Evolution of processes of heat generation along a flame in an OH furnace compared with the heat requirements*

Rational organization of heat-exchange and combustion processes

The performance (P) of an OH furnace can be defined as the mean value of the quotient obtained by dividing the effectively utilized heat by the total heat required, which is provided by fuel combustion ($\Delta\tau$kcal/ton).

$$P = \frac{(Q_H \cdot \eta)_m}{\Delta\tau} \text{ tons/h} \quad \ldots\ldots(1)$$

where Q_H is the heat input at the valve, kcal/h, $Q_H = H_e.B$; η is the coefficient of effective heat utilization, which is equal to the ratio of the effectively utilized heat to the heat of fuel supplied for the same period.

$$\eta = \frac{H_e + Q_m - Q_w - \frac{Q_e}{B}}{H_e} \quad \ldots\ldots(2)$$

where H_e is fuel combustion heat, kcal/m³ or kcal/kg; Q_m is the physical heat of the fuel and air before admission to the melting chamber, kcal/m³ of fuel or kcal/kg; Q_w is chemical and physical heat of the gases leaving the melting chamber in kcal/m³ of fuel or kcal/kg; Q_e is furnace chamber heat losses, kcal/h; B is fuel consumption, m³/h or kg/h.

Thus, to improve OH furnace performance, the mean value of the product $Q_H.\eta$ must be increased.

An increase in heat input (Q_H) depends on the presence of energy resources of the furnace, i.e. on gas and air pressure and on the draught and thus, in principle, it is easy to attain.

The value of the effective heat utilization coefficient (η) depends on the organization of the heat-exchange and combustion processes and, thus, control of η is difficult. When operating with air which is not enriched, $Q_w \cong H_e$, an increase in heat input without increasing η causes overheating of the outgoing end of the furnace and premature wear in this region. In view of what has been said, the possibility of increasing the mean heat input is associated in the closest possible way with the acceleration of the heat-exchange processes, while these are connected with the organization of the combustion processes and movement of gases.

Since, in OH furnaces, fuel is fed through one of the ports, there is a contradiction between accelerating the heat-exchange and combustion processes. To intensify the heat-exchange processes, heat extraction at the bath surface must be as uniform as possible. To increase heat input, the fuel combustion process must be intensified and this causes concentration of the maximum temperatures at the operating port as the temperature of gases leaving the melting chamber cannot be above 1600–1650°C. As a result, the higher the heat input, the less uniform is heat transfer at the bath surface.

TABLE I Relationship between furnace capacity and ratio of bath length to equivalent diameter of the outgoing section of the gas port

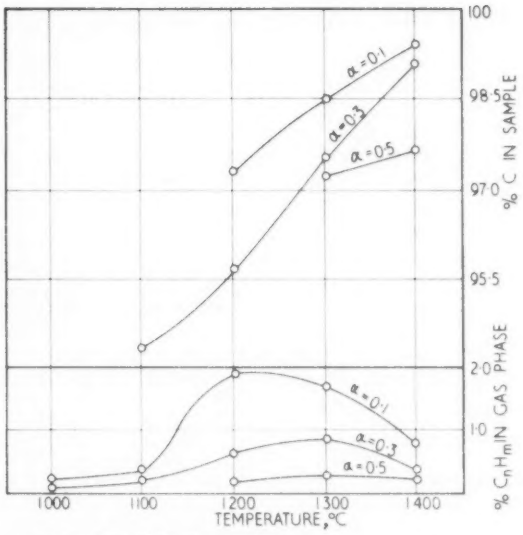
Furnace capacity, tons	35	85	125	200	500
$\frac{x}{(d_o)_{eq}}$	13·5	14·6	18·2	18·9	20·0

The curve in Fig.1 illustrates the heat evolution process in conformity with an exponential law, while the straight line shows the bath heat requirement under conditions of uniform heating.

The contradiction which has been mentioned is compensated to some extent by flame reversal, which is characteristic of the regenerative principle of the operation of OH furnaces, but the incongruity is not altogether removed and the greater the volume, and, consequently, the length, of a furnace melting chamber, the more telling it is. This is because, as furnace capacity is increased, the ratio of bath length (x) to the equivalent diameter $(d_o)_{eq}$ of the outgoing section of the gas port (Table I) increases.

A rational solution of this problem is the simultaneous operation of both ports and this can be realized in recuperative furnaces[32] or when converting gas-fired OH furnaces to fuel-oil firing or firing with high calorie gas,[33] and also by dividing the fuel supply through several burners through the roof.[34]

A simpler, though less radical method, is to construct ports in which flame length can be controlled within wide limits independently of heat input. The theory of this problem is briefly as follows. During the first stage of an OH heat (i.e. 30–40% of the total heat time in the scrap and ore process) the temperature of the surface of the refractories of the melting chamber is below the permissible limit while the difference in temperature between lining and charge surface is great. During this period, the lining plays an active part in heat exchange so that flame luminosity does not play any special part, while the process, as a whole, is limited by the heat generation (fuel combustion). For this reason, uniformity of heat transfer longitudinally in the bath is of no special importance during this period. In the second half of the heat (60–70% of the total heat time), the intensity of heat-exchange processes within the bath is limited by the properties of the refractories as their surface is at the limiting temperature. During this period, the uniformity of bath heating plays a decisive part while heat transfer must be organized on the principle of 'direct' directed heat exchange. The fundamentals of the theory of this view of heat exchange were stated in a report at a congress celebrating the Centenary of the French Société de l'Industrie Minérale.[29] During this period of the heat which is characterized by low heat inputs, the flame must cover the entire bath and possess an effective radiating capacity throughout its entire length. When fuel is supplied through one port, this can be achieved by high (degree of blackness up to 0·4–0·5) and as uniform as possible flame luminosity throughout the entire length of the bath while ensuring, at the same time, that there is no incomplete combustion in



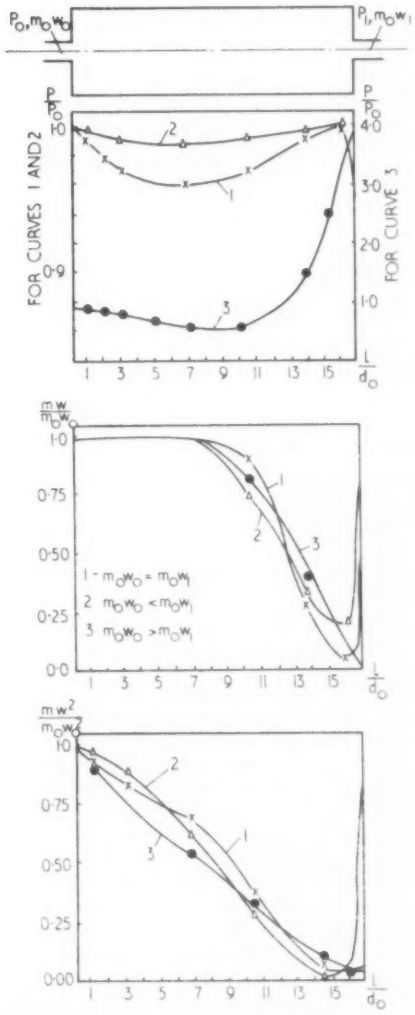
α coefficient of air consumption

2 *Variation in the amounts* C_nH_m *and* C *in relation to temperature in the reaction zone and to the composition of the initial gas–air mixture*

the melting chamber. It may be fairly confidently affirmed that ordinary lamp black (0·01–0·3 μ) gives unstable flame luminosity which, with intensive mixing, disappears in the region of the incoming port. To obtain stable flame luminosity in OH furnaces with dissociation of the hydrocarbons, slowly burning carbonaceous particles of large dimensions (1–3 μ) and of complex composition (the product of decomposition of hydrocarbon complexes which contains 5–6% hydrogen by weight) must be obtained. When burning liquid fuel, the higher carbon/hydrogen ratio in the fuel and its capacity for coking, the greater the probability of obtaining such particles (according to Conradson). As is known, when burning methane-containing gases, such particles are obtained if the decomposition of methane occurs at 1050–1150°C with the partial production of a liquid phase (tar).

Investigations carried out in the USSR show that the presence of a certain amount of oyxgen in a mixture of methane and air does not affect the composition of the decomposition products but it does displace the temperature range in which methane decomposition occurs with the production of liquid hydrocarbons toward higher temperatures (Fig.2). The complicated liquid hydrocarbon complexes which are formed during the decomposition of methane are gasified in the flame and subsequently dissociate with the separation of hydrogen, finally forming carbonaceous particles of complex composition. Such conditions are created also when burning diffused flames in which there are three-dimensional zones having the most varied methane/oxygen ratio.

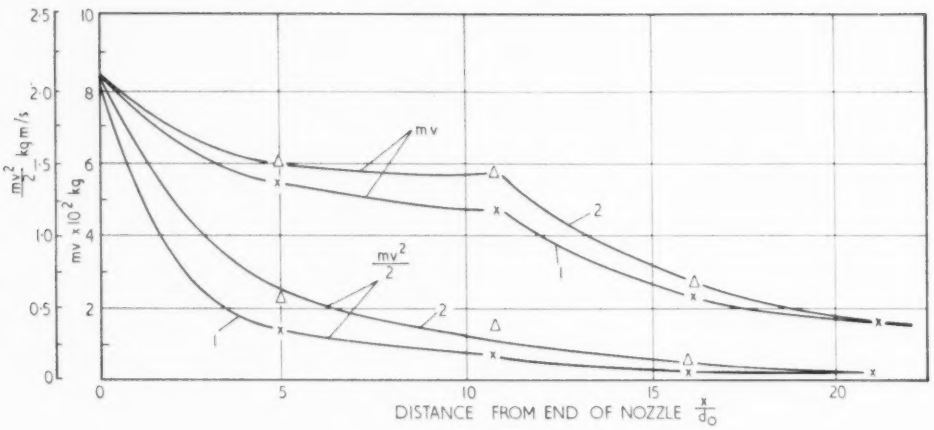
What has been said shows that versatility in the organization of the heat-exchange and combustion processes must be inherent in good OH furnace design. This versatility, in particular, is attained by the possibility of varying the angle of inclination of nozzles and burners which is possible in furnaces operating on liquid fuel and cold high-calorie gas. Furnaces operating on gas which is heated in regenerators are designed to create direct directed heat exchange and, thus, during the first period of the heat they do not operate under optimum conditions. This is, however, more obvious when operating with a solid charge, i.e. with the scrap process.



3 *Characteristic data for an axially symmetrical jet in a confined space*

Gas dynamics in the melting chamber

If a jet is directed along the axis of a hermetically sealed and confined cylinder, the recirculation of gases is organized under the dynamic action of this jet, the circulation zone being located between the jet and the walls. The greater the momentum (mw) of the jet, the greater is the degree of recirculation. When mw is constant, recirculation depends on the conditions of exit of the gases from the confined space. Figure 3 presents experimental data from the investigations of Glinkov and Choi En Hsir[30] showing the variation in pressure (axially), the momentum and the kinetic energy ($mw^2/2$) of a jet in a confined space. Since the movement of the counter-flow stream along the walls is ensured



Curve 1 for $\beta = 0°$
Curve 2 for $\beta = 10°$

4 *Characteristic data for a jet moving along a hearth*

by the difference in pressures between the end and the beginning of the space, maximum recirculation evidently corresponds to curve 3 ($w_0 > w_1$) and the minimum to curve 2 ($w_0 < w_1$). A similar conclusion can be reached if we compare the course of the relevant curves for mw and $mw^2/2$. Thus, the less the kinetic energy of the outgoing stream compared with the initial energy, the more energetic is the recirculation of the gases. Approximate constancy of momentum in the first half of the trial chamber is not a general rule but refers only to a chamber of a certain transverse dimension. If, in this same chamber, a jet is arranged at the hearth and directed parallel to it, a reduction in the momentum of the jet begins to be reflected gradually at the exit of the jet from the aperture. This is illustrated by the upper curves in Fig.4 in accordance with the investigations of Mitkalinny and Ni Hsueh Chu.[25] Curves '2' in (Fig.4) refer to the case where the jet is directed at an angle of 10° to the hearth. The greater the cross-sectional area of the confined space, the greater the expansion of the jet and the greater the proportion of the initial kinetic energy used to overcome the resistances of the movement of the jet in the confined space (i.e. the more energetic the recirculation), and the less is the amount transformed to potential energy (pressure). For a hot jet, this conclusion is only valid if the effect of heat generation in the flame on gas mechanics can be taken into account, making it possible to calculate the thermal and hydraulic sides of the process independently of each other.

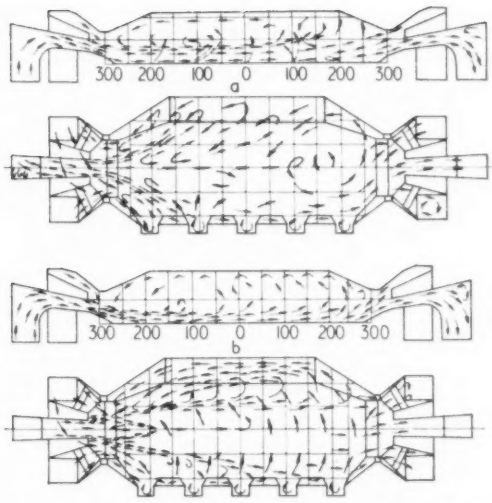
As is known, the kinetic energy of gas and air streams in the OH furnace is usually between two and two-and-a-half times as great as the kinetic energy of the combustion products at the exit. The less the infiltration of atmospheric air and gas evolution from the bath, the greater this difference. Thus, conditions are created which are characterized by curve 3 (Fig.3) i.e. more favourable for the creation of recirculation under the roof. The presence of this recirculation, if it is fed by the outgoing gases which are cooler than the roof, has a favourable effect on roof life. Thus, an increase in the energy factor $f_E = \frac{m_0 w_0^2}{m_1 w_1^2}$ in OH furnaces is useful from the point of view of developing recirculation. When the stream is directed at the bath at an angle, maximum flame velocity is displaced to the bath surface, [29] the resistance to friction increases, and this causes a certain reduction in the capacity of the flame to create recirculation.

Table II gives the energy factors for various fuels, based on the condition that the infiltration of atmospheric air is compensated for by the gases being driven out.

From the above mentioned considerations, it follows that, from the point of view of gas recirculation in the melting chamber, other things being equal it is not the absolute value of the kinetic energy of gas and air which is of special importance but their relationship to the kinetic energy of the combustion products at the exit from the melting chamber (f_e). Physically this can be explained by the fact that, with a positive value of the difference in the kinetic energy at the inlet and

TABLE II Energy factors for various fuels

Fuel	Combustion heat, H_c, kcal/m³	Gas velocity in port at 0°C, m/s	Air atomizer velocity at 0°C, m/s	Air velocity in port at 0°C, m/s	Energy factor, f_e	Difference in kinetic energy at inlet and outlet, ΔE, kgm/s	Ratio of fuel mass to air mass, M_t/M_a
Mixed gas	1800	10	—	2	2·45	+0·595	0·45
Mixed gas	1800	8	—	2	1·72	+0·42	
Mixed gas	2500	10	—	2	2·44	+0·59	
Mixed gas	2500	8	—	2	1·7	+0·41	
Mixed gas	2500	10	—	4	0·90	−0·114	
Natural gas	7000	100	—	2	4·13	+0·83	0·06
Natural gas	7000	100	—	4	1·35	+0·258	
Fuel oil	9700 kcal/kg	—	200	2	14·70	+0·93	0·07 0·20 taking into account atomization by air



5 *A pattern of gas movement in the melting chamber of a 185-ton gas-fired OH furnace operating on* (a) *air alone and* (b) *enriched air*

outlet, there is excess kinetic energy (and, thus, an excess of momentum) which can be used for recirculation.

$$\Delta E = \frac{m_0 w_0^2}{2} - \frac{m_1 w_1^2}{2};\ (m_1 = m_0) \quad \ldots\ldots(3)$$

With a negative value of ΔE and with inadequate draught, the stream, throughout its entire extent, must possess great potential energy essential for creating a suitable velocity in the outlet section, and which cannot be used for recirculation.

Table II also shows that an increase in the velocity of the fuel stream increases f_e; on the other hand, an increase in the velocity of the air decreases f_e, i.e. reduces recirculation.

From what has been stated, it follows that an increase in the velocity of the outgoing gas stream caused by feeding compressed air into the gas port or by supplying oxygen or superheated steam through the port in independent jets extends the recirculation zone and intensifies recirculation.

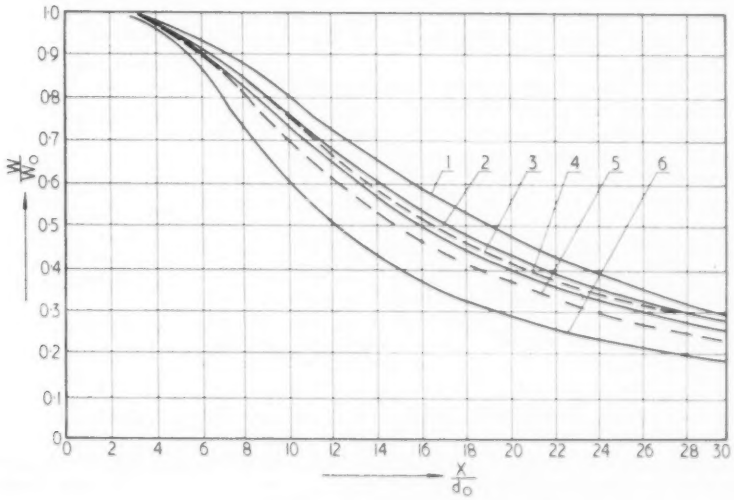
Based on data obtained by Ivantsov and Sobakin, on the model of a 185-ton gas-fired furnace, Fig.5 presents a flow pattern for operations (*a*) without oxygen and (*b*) for feeding oxygen through two pipes located at the sides of the gas port at an angle of 7·5° at a rate corresponding to an enrichment of 40%. As the investigations showed, such an oxygen supply to the flame is more rational for venturi ports and is at present being used in many Soviet plants.

The main question associated with gas dynamics in the furnace chamber is that of flame bath-covering capacity, i.e. capacity for maintaining proximity to the bath surface throughout the entire extent of the melting chamber. From a physical point of view, the nature of flame bath-covering capacity arises from the fact that streams flowing along the surface suffer less frictional drag so that flow along the surface spreads further (long-range flame). Figure 6 gives experimental data obtained by Maslovsky[14] on the long-range capacity of jets flowing along the surface, for various shapes of nozzle cross-section, in comparison with a free jet for which the cross-sectional shape of the nozzle does not affect the long-range capacity.

As the curves in Fig.6 show, the long-range capacity of a bath-covering jet depends on the value $K = \frac{P_f}{P_\Sigma}$ where P_f is the free portion of the perimeter of the nozzle, and P_Σ is the complete perimeter of the nozzle. The lower the value of K, the greater the long-range capacity of the jet. If the stream is directed at the surface at an angle (β), maximum velocity in the stream is displaced to the surface.[29] This improves the bath-covering capacity of the flame although, at the same time, the stream spills on to the surface at an angle of β_r greater than for the case of $\beta = 0$, which reduces the long-range capacity. According to experimental data[28]

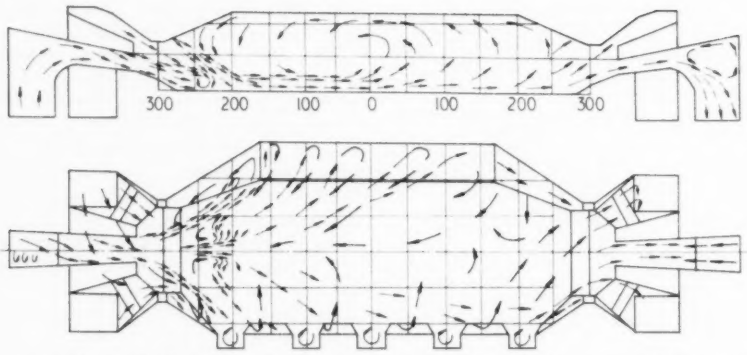
$$\beta_r = 30 + 3 \cdot \beta \quad \ldots\ldots(4)$$

Thus, generally speaking, there is an optimum angle of incidence β, corresponding to maximum long-range capacity. When a jet flows in a space which is confined



1 semi-circular nozzle K = 0·63;
2 triangle K = 0·667;
3 rectangular K = 0·7;
4 square K = 0·75;
5 rectangular K = 0·8;
6 a free jet

6 *Long-range jet flowing along a wall*



7 *Pattern of the movement of gases in the melting chamber of a 185-ton gas-fired OH furnace working on enriched air*

by walls, if the flame is spread at an angle β_r, the greater β, the more rapidly does the stream begin to flow at the longitudinal walls. According to the experimental data of Mitkalinny,[24] the reduction in velocity in an isothermal jet spread closely along the wall in a confined sapce is determined by the expression

$$\frac{w}{w_0} = e^{-a\left(\frac{x}{d_0}-4{\cdot}5\right)} \quad \ldots\ldots(5)$$

where a is a constant coefficient.

An investigation of the long-range capacity of streams in a model of a gas-fired OH furnace with a venturi port, as compared with flow of similar streams in an unconfirmed space, showed that confined streams at first $\left(\text{up to } \frac{x}{d_0}=18\right)$ decay more slowly but subsequently, more rapidly than unconfined streams. In this way, the greater the air consumption/gas consumption ratio, the greater is the long-range capacity of the confined stream. Branches of the stream at the wall rise and, subsequently, join the recirculation under the roof. The streams mentioned are undesirable because they form the high-temperature portion of the flame and cause overheating of the walls and roof, giving rise to erosion as they also entrain slag dust. Figure 7 shows a flow pattern in the melting chamber when the oxygen jet is directed from the roof to the mouth of the port flame aperture at an angle of 41°. This causes marked deterioration in bath-covering capacity.

Investigations[31] show that if nozzles or burners are directed at each other at an angle to the horizontal plane, the jet which is formed after impact flows at the wall to a lesser extent. Thus, for burners inclined to the bath at an angle of 12°, the optimum horizontal angle appeared to be 30°. It is not difficult to show by fairly simple calculations that in the case of isothermal flow there is no great difficulty in obtaining a flame which covers the entire bath. In full-size furnaces, however, the flow of a flame formed by gas and air streams is not isothermal: the initial temperature of gas and air streams is 1100–1200°C, flame temperature varying from 2200–1650°C. The gases outside the flame in the melting chamber have a temperature of 1600–1800°C. As a result, considerable lifting forces act on the flame. In addition, a vertical stream of carbon monoxide and gases liberated from the bath also try to deflect the flame from the bath surface.[11]

The equation used to determine the vertical deflection of the axis of a bath-covering stream by the action of gravitational forces and of the gas stream from the bath has the form

$$\bar{y} = \frac{K_3}{0{\cdot}82.a_1} A_r \sqrt{\bar{x}} + 0{\cdot}4\ [0{\cdot}4\bar{x}^2 + 0{\cdot}053\ (\bar{x}-0{\cdot}08)] + 0{\cdot}83\frac{W_g}{W_a}\sqrt{(\bar{x}+0{\cdot}4)^3} \quad \ldots\ldots(6)$$

$$\bar{y} = \frac{a_1.y}{c_0};\ \bar{x} = \frac{a_1.x}{c_0};\ A_r = \frac{gc_0}{W_0^2}.\frac{T_0-T_m}{T_m}$$

where W_g are exit velocity of CO from the bath in m/s; W_0, T_0, c_0 are, respectively, the actual velocity, absolute temperature, and half the thickness of a bath-covering jet at the point of its formation after impact of the flame against the bath surface; x and y are coordinates of the point under examination along the axis of the flame, m; T_m are absolute temperature of the medium surrounding the flame; K_3 and a_1 are constant coefficients ($K_3=0{\cdot}7$ and $a_1=0{\cdot}1$).

Table III gives data for calculating the value of 'y', for values $T_0=2000$°C; $T_m=1700$°C; $c_0=0{\cdot}75$ m, and $W_g=0{\cdot}3$ m/s. The data in Table III indicate the relationship between the bath-covering capacity of the flame and gravitational forces. The value $W_0=10$ m/s corresponds to an outlet velocity in the gas port of 55–60 m/s, characteristic of large furnaces fired with mixed gas and subject to maximum heat input.

The parameter $x=10$ m corresponds to furnaces of 125–150 tons capacity. When furnace capacity is increased, and consequently parameter x, the bath-covering capacity of the flame in the second half of the melting chamber deteriorates considerably.

Gas statics in the melting chamber

The complex nature of the movement of gases in a confined space determines the existence of a no less complex field of static pressures which is found as the result of superimposing three existing fields,

$$P(x, y, z) = P_l + P_t + P_w \quad \ldots\ldots(7)$$

where P_l is the field of pressures created by external factors, for instance, atmosphere, the chimney, the exhaust fan, etc., and which, according to Pascal's law, are distributed in a similar manner at all points within a confined space. This component can be characterized as a pressure level; $P_t = f(x, y, z, t)$ = the field of pressures created by the geometrical pressure head and

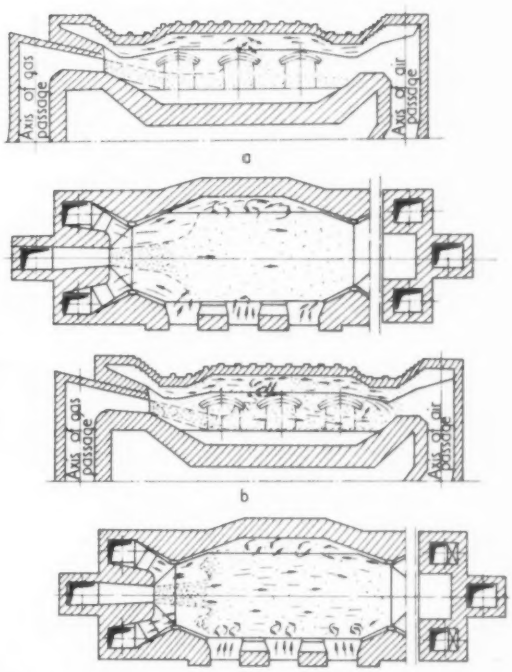
TABLE III Relationship between the bath-covering capacity of the flame and gravitational forces

m/s	Value of '*y*', m 10	12	15	16
5	0·43	0·63	0·97	1·5
10	0·14	0·2	0·29	0·44
15	0·07	0·1	0·15	0·22

depending on the temperature and physical properties of the gas. For investigations on cold models, $f(x, y, z, t)=0$, of course. $P_w=f(x, y, z, w)$ is the field of pressures arising as a result of gas dynamics.

The geometrical pressure head gives the component of the field of pressures. This component increases in value from bottom to top. The component of the field of pressures, as a result of gas dynamics, depends on static pressure in the jet, and in the circulation zones. In a hermetically sealed space, the value P_w along the flame first begins to fall and subsequently increases. The greater f_E the greater the extent of this increase. On the other hand, with low values of f_E (ΔE a negative value) the pressure P_w is practically constant along the length of the flame. If, in the outlet section, the pressure is reduced, then the curve of pressure P will be displaced parallel to itself. In OH furnaces, because of the presence of several relatively non-airtight doors, the pressure level is mainly determined by atmospheric pressure. When the effect of draught is extended to the melting chamber, atmospheric air inevitably infiltrates in intolerable quantities.

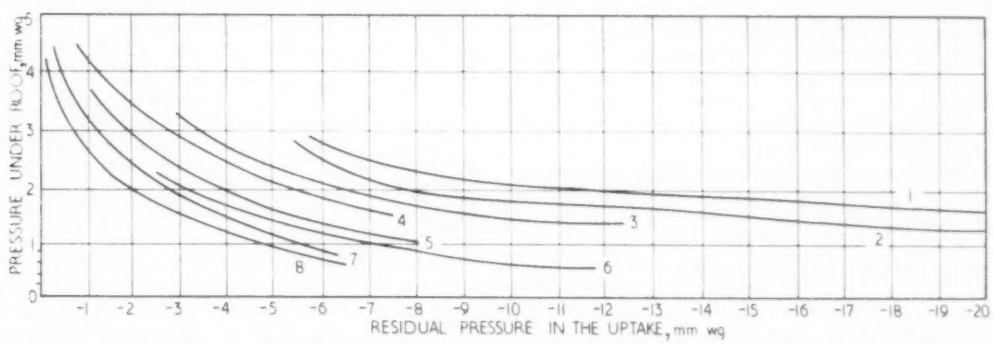
For the reason indicated, the pressure level (P_l) in the furnace chamber cannot vary arbitrarily and, thus, the effect of the chimney (or exhaust fan) draught is limited. This is illustrated by the curves in Fig.8 based on measurements made on full-size furnaces. As we see, only with a clearly inadequate draught and as a result of upsetting gas dynamics in the melting chamber, does a reduction in draught lead to some increase in pressure under the roof. This leads to the important conclusion that the location of the so-called zero surface $P(x, y, z)=0$, depends on f_E and must be the most rational for any given furnace. This follows from the optimum combination of losses with the exhaust gases and as a result of the infiltration of atmospheric air. If, with inadequate draught, the stream must dispose



a normal furnace operation
b furnace operation with reduced draught

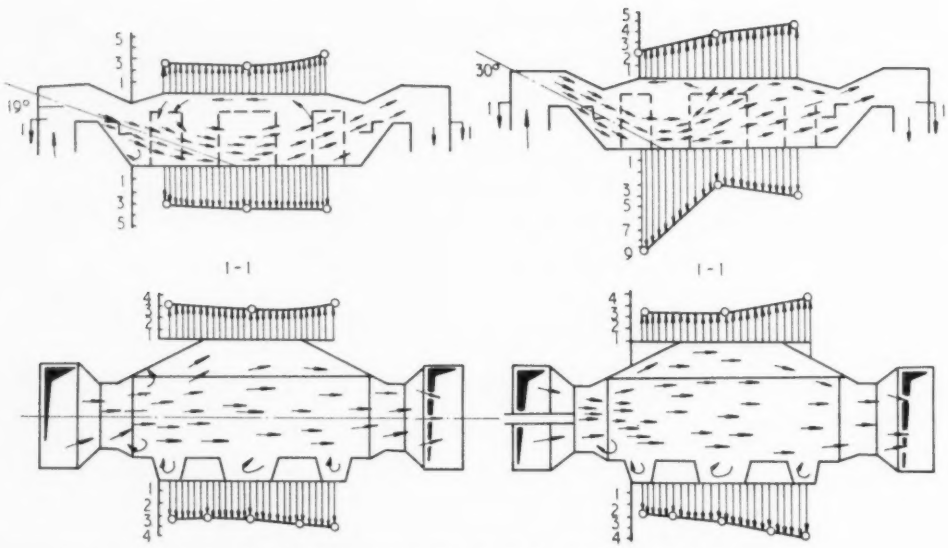
9 *Movement of streams of gas, air, and combustion products in the melting chamber of a 35-ton OH furnace*

of an excess of static pressure before reaching the outgoing section, which will ensure an increase in the velocity of the stream up to a value characteristic of the outgoing section, then, at values of $f_E<1$, this will increase the value of P throughout the entire length of the flame (Fig.3). The latter causes intensified exhaustion of gases and upsets the dynamics of gas movement in the melting chamber because of the greater separation of the gases from the flame in the charging door apertures. Although, in a hermetically-sealed space, variation in draught does not influence gas dynamics but affects only P_l, in the melting



1 and 2 380-ton tilting furnaces V_w is equal to $53\cdot10^3$ and $46\cdot10^3$ m³/h
3 and 6 380-ton tilting furnaces V_w is equal to $50\cdot10^3$ and $34\cdot10^3$ m³/h
4 300-ton fixed furnaces $V_w=50\cdot10^3$ m³/h
5 190-ton fixed furnaces $V_w=34\cdot10^3$ m³/h
7 380-ton tilting furnaces $V_w=40\cdot10^3$ m³/h
8 190-ton fixed furnaces $V_w=30\cdot10^3$ m³/h

8 *Relationship between pressure under the roof and residual pressure in the uptakes for furnaces with different amounts of the combustion products* (V_w)



10 *Pressures (mm wg) in a longitudinal section and in a plan of a model of a 40-ton OH furnace*

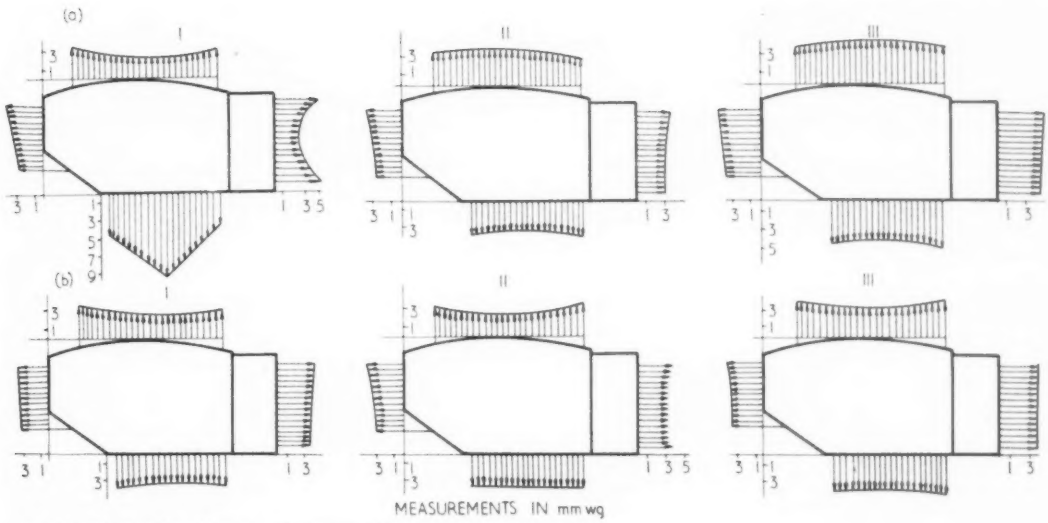
chamber of an OH furnace a variation in draught has little effect on P_l but can considerably affect the pressure in the zone of the flame and thus has a substantial effect on gas dynamics. With high values of f_E the influence of inadequate draught is less apparent since the natural increase of pressure (P_w) toward the end of the flame in itself partly compensates for the lack of draught. On the other hand, when there is a reserve of draught it is possible to ensure an increase in the velocity of the stream up to the value obtaining in the outlet section and, thereby, avoid an increase of pressure in the zone of the flame even with low values of f_E.

This is well illustrated by the investigations of Gol'denberg[17] on an actual OH furnace. When draught was reduced by means of reducing the cross-sectional area of both outgoing air uptakes by one third, the flame, even in the first half of the melting chamber, was less organized, the flame boundaries were sharply curved, additional turbulent regions were created, and exhaust of the gases from the doors was intensified (Fig.9). Although a reduction in draught can be obtained by completely covering one of the air uptakes symmetry of movement of the gases is disturbed. A very complex relationship between the field of static pressures in the melting chamber and the factors mentioned above make it extremely difficult to create a rational scheme of automatic pressure control.

For OH furnaces, the question of gas pressure at the roof and wall is of great importance. Observations show that the wear of the refractories is greater at places of high velocities and when gases filter through the lining. This is because the gases contain a large amount of slag dust. On the other hand, an increase in pressure on the bath improves heat transfer. Figures 10 and 11 present data from research carried out by Glinkov and Rekhtman[16] on the pressure of gases on the walls and roof of an OH furnace. These data were verified by measurements effected under production conditions. They indicate the relationship between pressure on the walls and roof, and the gas dynamics which, in turn, are governed by the direction of the stream from the nozzle. As Figs.10 and 11 show, lining wear at the end of the melting chamber must be greater than at the beginning, in which case with an angle of inclination of the nozzle of 30°, wear will be less uniform. These data correspond with research data[26] on dust precipitation on a model of a 30-ton OH furnace. Coarse dust was modelled by bubbling air through a water bath, and fine dust by feeding water vapour through the bath surface. When studying dust precipitation, the direction of the movement of the gases was periodically changed (reversing). Figure 12 shows that coarse dust (from 0·2 to 1 mm) settles more at the ends of the melting chamber where high pressure obtains, and fine dust (less than 0·1 mm) is deposited practically uniformly and independently of pressure conditions.

Mixing of gas and air streams and flame structure

Processes associated with the mixing of gas and air streams were the subject of numerous investigations in the USSR using cylindrical rigs and apparatus simulating the melting chamber of an OH furnace. The conditions of mixing cold streams and of a gas–air stream in a hot flame are not similar, although, in so far as the combustion processes are limited by diffusion processes of mixture formation, a study of the process of mixing in isothermal models provides very important material for evaluating combustion processes in actual furnaces. The overall effect resulting from an analysis of the distribution of surfaces of equal concentrations shows that the mixing process is accelerated when gas and air velocities are raised and the angle of incidence (within known limits) of these streams is increased. From the point of view of flame organization, especially as regards flame symmetry in the melting chamber, the distribution of air at the perimeter of the gas jet is of great importance. The graphic method of evaluation of the distribution of air suggested by Belov and Malysheva[21] involves the con-



a angle of inclination of burner 30° *b* same, 19°

11 *Pressures (mm wg) in cross-sectional areas of a model (at the doors)*

struction, for various sections of a gas–air stream, of a hodograph of the vector ($a'\phi$) characterizing the inflow of air into a gas jet along a straight line ϕ proceeding from the centre of the jet relative to the total inflow of air through the contour of this jet and of a hodograph of the vector ($a\tau$) characterizing the relative amount of air flowing in along a straight line ϕ through the contour of the air stream.

Evidently

$$a'\phi = \frac{100}{(Va)i_0}\int^{r_0} W_a.r.dr \ \%$$

$$a\phi = \frac{100}{Va'_0}\int^{R_0} W_g.r.dr \ \%$$

where Va and Va' are, respectively, the air consumption per second through a model and the amount of air which has flowed into the gas jet as far as the secton in question f_i

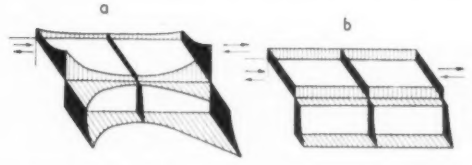
$$(Va)_i = \int_{f_i} Wa^{df_i}$$

W_g and W_a are, respectively, the 'delivery' velocities of the air and gas streams obtained by multiplying the resultant velocity of the stream by the concentration of gas and air.

r_0 and R_0 are the radii of the contours of gas and air streams respectively in the direction ϕ.

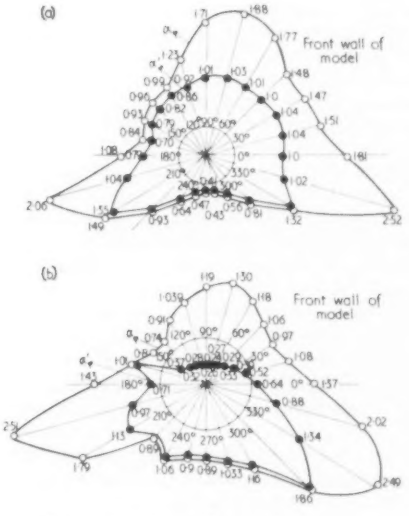
Figure 13 presents the hodographs of the vectors $a'\phi$ and $a\phi$ for two sections of a model of a melting chamber of a 185-ton OH furnace with venturi ports. The nature of the hodographs of the first section quite clearly indicates the non-uniformity of the inflow of air into the gas stream and the inadequacy of the inflow of air from the bath side. On the other hand, in the second section, the inflow of air from the bath side, especially in places adjacent to the front wall, is the most intensive. This is all the more important because the oxygen of the gas phase, in addition to ensuring the process of combustion, has the very important technological functions of ensuring oxidation of the bath impurities.

Until recently, investigations in the field of aerodynamics of OH furnaces referred mainly to hydrodynamic phenomena and were characterized by visual



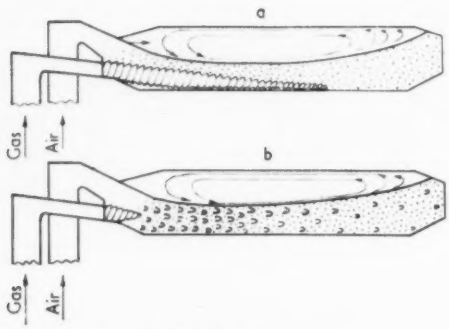
a coarse dust *b* fine dust

12 *Pattern of dust deposition on the walls and roof of a model*



a section $\frac{x}{d_0} = 3{\cdot}42$ *b* section $\frac{x}{d_0} = 5{\cdot}6$

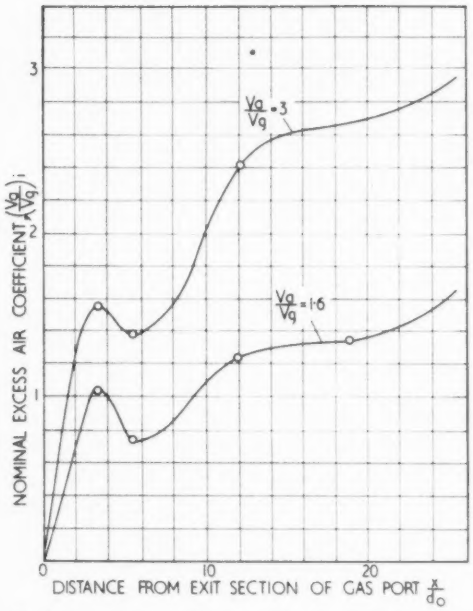
13 *Picture of the admission of air into a gas stream for* $\frac{Va}{Vg} = 1{\cdot}6$



a with a low scale of turbulence
b with a high scale of turbulence

14 *Flame structure in the absence of turbulent masses of the second category*

observations and a study of the fields of the average values (speeds, pressures, and concentrations) which do not provide an adequate picture of actual flame structure at each instant of time and of the mechanism of the processes of mass exchange in the flame. According to the theory of a hot flame,[28] as distinct from a laminar flame, the turbulent flame obtained by burning gas in an atmosphere of air, has a complex structure. The continuous combustion front, natural to a laminar flame and which divides the regions of pure gas and the mixture of the combustion products with air, is preserved in a turbulent flame until the scale of turbulence (l) is less than the width of the flame front. When the scale of turbulence exceeds this value, the phenomenon of rupturing of the front occurs and, consequently, the front no longer actually exists as a stable shape but is constantly disturbed by the intrusion of turbulent masses of air and the carrying away of turbulent masses of gas as a result of turbulent pulsations. The phenomenon of volumetric combustion occurs everywhere where, according to the law of probability, stoichiometric relationship between gas and air is attained. Naturally, at the beginning of the flame the probability of achieving a stoichiometric relationship is very great and, consequently, heat generation occurs intensively. On the other hand, at the end of the flame where the volumes of unburnt gas are scattered among the mass of combustion products and excess air, this probability is relatively slight. This also gives rise to the phenomenon of incomplete combustion, generally speaking, not natural to a laminar flame, the combustion front of which is completely defined. According to the theory of turbulence, the greater the pulsating component of velocity (W'), the greater the scale of turbulence; this component in turn depends on the average velocity of the stream. The product lW' is characteristic of the intensity of a turbulent mixture. The turbulence of a free jet degenerates, i.e. pulsations with great degrees of turbulence are gradually converted to pulsations with a lesser degree. However, when a jet flows in a confined space and the streams collide with solid and liquid surfaces and with each other, pulsations with large-scale turbulence again occur between them. The investigation of flame structure in OH furnaces is practically only beginning but even the first work in the USSR in this field[19, 23, 35] indicated the practical importance of this side of the question. The following concepts must

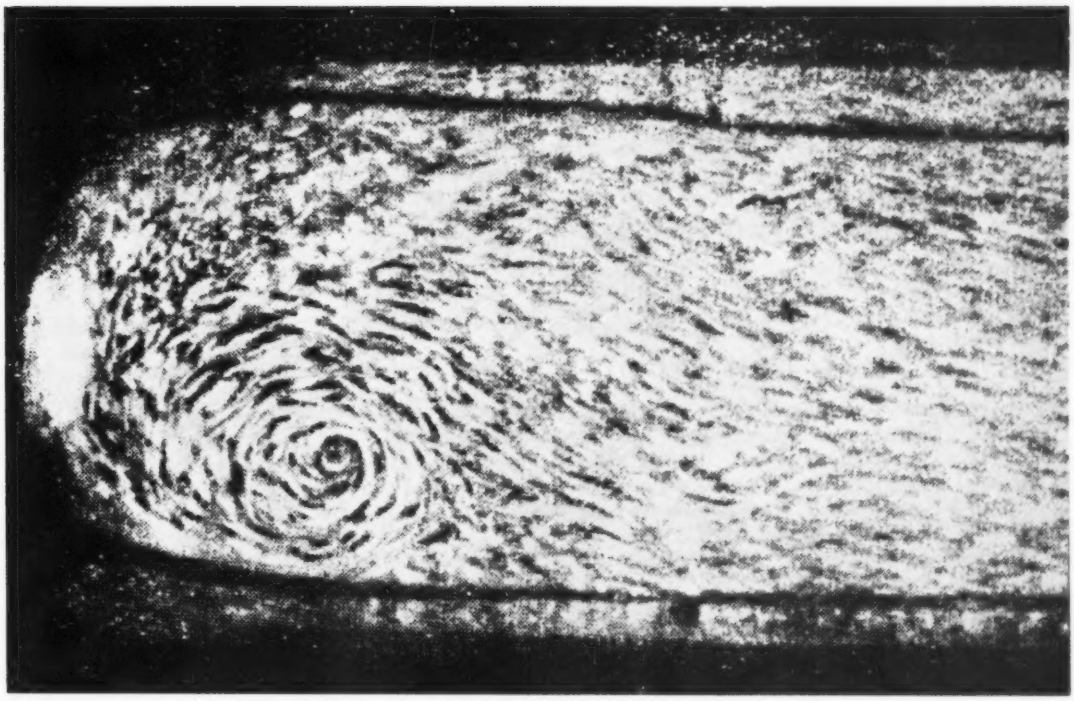


15 *Variation in nominal excess air coefficient in the flame* $\frac{(Va)i}{Vg}$ *along a model of a 185-ton gas-fired OH furnace*

still be considered as no more than a working hypothesis demanding further experimental verification. In the presence of a continuous front of combustion (a low-scale turbulence in the stream) the process of heat generation occurs relatively slowly but in the combustion products behind the front there are practically no unburnt portions of gas. The greater the scale of turbulence, the more intensive is the fuel combustion process but in the combustion products there are volumes of unburnt gas and, thus, incomplete combustion (Fig.14).

The two categories of turbulent masses which constitute the stream of a flame must be distinguished. The first category embraces turbulent masses, the size of which is governed by a combination of various degrees of turbulence of the inlet gas and air streams and depends on stream velocities. When two streams mix, this category of turbulent masses is more or less uniformly distributed among the streams and determines the velocity of turbulent diffusion and thereby rapidity of fuel combustion. Investigations by Lisienko, Kokarev, and Troib[35] using cine-photography on flames at a speed of 220–240 frames/s showed that the frequency of pulsations of turbulent fuel-oil flames exceeds the lower limit of sound frequency (40–60 pulsations/s) and, with distance along the furnace, it is reduced to two-fifths to a third. The velocity of movement of vortexes varies from 40–60 m/s at the beginning of the melting chamber to 30–40 m/s at the end of the chamber. The greater the atomizer pressure, the higher the frequency of pulsations and movement velocity of the flame vortexes. The high frequency of pulsations of a fuel-oil flame is evidently due to a sonic effect characteristic of a good fuel-oil flame.

The second category of turbulent masses, generally speaking, is not characteristic of streams and arises in



16 *Turbulency of a stream on admission to the gas port of a model*

them on contact with solid surfaces, for instance, with flow around obstacles (bends and sudden widening in the channels of the ports; in pipe lines, impact against the charge in the melting chamber, etc), when the flame streams collide with the uneven surface of the liquid bath or when the streams collide with each other (gas and air at large angles of incidence, when the flame stream collides with the gases liberated from the bath, etc.).

Although turbulent masses of the first category are unavoidable with intensive fuel combustion, turbulent masses of the second category are extremely undesirable since they cause local disturbances of stream structure and, thereby, incomplete combustion and a deterioration in conditions of direct directed heat exchange. To discover a means of overcoming this phenomenon is the main task in OH furnace design. The greater the kinetic energy of the main stream (i.e. velocity) and thus its rigidity, the less probable it is that it will be destroyed by turbulent masses of the second category. It is for this reason that the distorting influence of these turbulent masses is so great in the second half of the length of the melting chamber where the velocity in the stream is relatively less. On the other hand, conditions which favour the formation of turbulent masses of the second category in the ports must be eliminated by creating good streamlining, uniform air distribution in the uptakes, and by straightening out the stream in the gas port.[29] In view of this, substantial incomplete combustion which sometimes occurs with high thermal inputs, is evidently the result of disturbing the flame structure. The difficulty of completely burning the gas residues is aggravated by recirculation of the gases since admission of the return combustion products to the main stream impoverishes it in relation to the amount of oxygen. As regards the excess-air coefficient in the contour of the gas stream, its variation along the flame is very complex and evidently not very stable (Fig.15)[21] due to the exchange of turbulent masses. However, it appears that it is, in fact, the complex turbulent structure of the flame which ensures the energetic admission of oxygen to the bath for oxidizing the impurities and, consequently, the oxidizing capacity of the furnace increases as turbulent mixing in the flame is intensified.

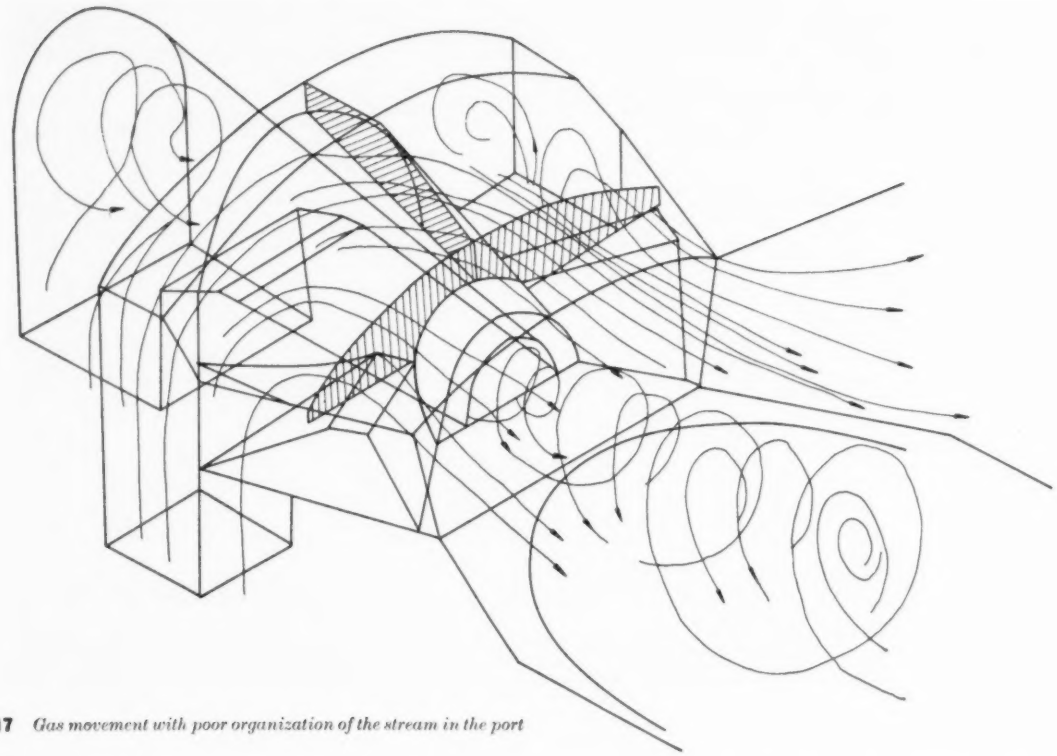
Port design in OH furnaces

Ports are the structural elements of a furnace which ensure conditions of admission of gas and air into the melting chamber in a more favourable manner from the point of view of promoting combustion and heat exchange and, consequently, the following points arising from the above analysis must be taken into account:

(i) the streams of gas and air leaving the ports must be controlled and have an axial direction and, so far as possible, turbulent masses of the second category must not occur in them

(ii) one of the streams (predominantly a fuel stream) must be the main energy carrier because this facilitates the creation of a more stable organization of a bath-covering flame and the attainment of effective direct directed heat exchange

(iii) an increase in the velocity of the fuel stream (the main energy carrier) accelerates the processes of mixing and heat exchange, intensifies recirculation, and makes it possible to



17 *Gas movement with poor organization of the stream in the port*

operate with smaller draught reserves. The pressure drop along the flame is increased

(iv) an increase in air stream velocity accelerates the mixing processes but retards recirculation and also causes an increased pressure along the flame. Consequently, an increase in velocity of air preheated in the regenerators is desirable within known limits (up to $f_E > 1$); a further increase in air velocity (with inadequate draught) can lead to deterioration in flame organization

(v) the non-uniform admission of air through the air port disturbs the organization of combustion and flame structure and the greater the proportion of energy which is carried by the air current, the greater the extent to which this occurs

(vi) the admission of an adequate amount of oxygen into the zone of the flame adjacent to the bath surface plays a most important part, both in the organization of heat exchange and in accelerating the progress of technological reactions (oxidation of the impurities)

(vii) an increase in gas recirculation (within known limits) facilitates conditions associated with lining wear and especially roof wear but it causes a deterioration in conditions associated with complete burning of fuel residues at the end of the flame

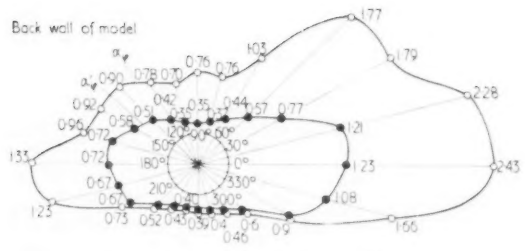
(viii) the influence of draught in OH furnaces is reflected mainly in the conditions of static pressures in the flame but not in the level of pressure in the melting chamber (P_l), and consequently inadequacy of draught has an adverse effect on flame organization and adversely affects heat-exchange conditions.

The organization of the gas stream in the port is of special importance in hot-gas fired furnaces. Uniform admission of gas from the uptake into the gas port, a gradual taper of this port with an adequate, even surface, and high exit velocity, create conditions for obtaining a well organized flame having an axial direction.

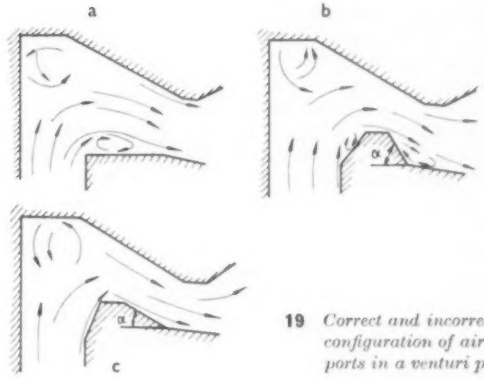
Figure 16 shows photographs of a vortex formed during the entry into the gas port. As a result of the presence of this vortex, the gas stream in the gas port is twisted or curled as a result of which a turbulent, rather unstable flame is obtained and the fuel is only partly burnt (Fig.17).

Experiments on models and actual furnaces with Siemens and venturi ports have definitely established that non-uniform admission of air through the uptakes leads to non-uniform admission of air into the gas stream and upsets the organization of the combustion processes in the flame, also leading to poor utilization of the air and to incomplete combustion.

Figure 18 shows hodographs of vectors $a'\phi$ and $a\phi$ with admission through the back wall uptake of twice the amount of air. In this case, the air stream tends more strongly toward the front wall and creates the impression that the gas stream is forced toward the back wall. When all the air is admitted through one of the uptakes, a well developed recirculation zone can even be obtained at the longitudinal wall of the melting chamber from the same side.[16,29] Actually, the forcing back of the gas stream is only a visual impression since only the position of the more visible zone of the flame is changed and this depends on the proper-



18 *Admission of air into a gas stream for* $\frac{Va}{Vg}=2{\cdot}0$ *for section* $\frac{x}{d_0}=3{\cdot}42$



19 *Correct and incorrect configuration of air ports in a venturi port*

ties of the fuel and the distribution of the air. Thus, when feeding air in a large quantity through the front uptake, less air will be admitted to the flame from the direction of the front wall, the flame here is less transparent and, consequently, the impression is created that the gas stream is forced back to the front wall.[17] In all designs of ports with two air uptakes, the difficulty in obtaining uniform distribution of the air in the uptakes involves a variation in the conditions of distribution as the slag pockets are filled and, consequently, a constant variation in the cross-sections of the uptakes does not remove completely the non-uniform distribution. Thus, the advantage of a single uptake port is to be found not only in the simplicity of its design and the ease of repairs, but also in the greater possibility of ensuring uniform admission of air into the melting chamber. It is possible that the better streamlining of the internal cavities of the air passage of the port plays a great part. Consequently, the shape of the air ports of the venturi port shown in Fig.19*c*, as the investigations show, is more fortunate in relation to the formation of turbulent masses of the second category than the one shown in Figs.19*a* and *b*. It should be pointed out that conditions of streamlining usually deteriorate substantially as the ports begin to burn. To consider the question of the angle of incidence of air and fuel streams and the part played by air velocity, we must consider the effect on the combustion processes of a relative mass of fuel (M_f/M_a) the ratio of mass of fuel to air, depending on the calorific value of the fuel.

Table II gives this characteristic for three types of fuel. These data show that if we wish to create identical energy conditions for the formation of a mixture when burning mixed gas, atomized fuel oil, and natural gas, the exit velocities of the gas must be in the proportion of 1:2·25:7·5, respectively, for a given air velocity. When burning fuel oil, in particular, if burners with a Laval nozzle are used, substantially higher fuel stream velocities are attained than are provided for by the proportions indicated above and, consequently, for this type of fuel the part played by the angle of incidence of air and fuel streams is practically negligible. For the same reason, air exit velocity can also be low. When burning natural gas, the part played by direction and velocities of the air stream is negligible only when burner exit velocities exceed 450 m/s. When burning producer and mixed gases, the higher the velocity of the fuel stream, the less may be the angle of incidence of the air and fuel streams.

In practice, port design in the USSR in recent times has, to a great extent, involved an increase in the energy level of the fuel stream or of the gas–air mixture by feeding air, steam, or oxygen at high velocity. Each of these measures is effective but in different ways. The effect of feeding compressed or fan (turbine) air into the gas stream of the port is obtained by burning part of the gas in the gas port, as a result of which there is an increase in the velocity and kinetic energy of the gas stream. With the introduction of 10% of the theoretically necessary air into the gas port the temperature of the mixed gas (2000 kcal/m^3) increases by about 400°C and, probably, also the kinetic energy of the stream at the exit is raised by about 100%. The velocity of air feed to the gas port is operative here only as a factor of mixture formation and also from the point of view of overcoming the resistance of the gas port. To obtain a similar effect when feeding the same 10% of air directly into the melting chamber, it should be fed at a rate of 115–120 m/s. If air is fed at the rate of about 250 m/s, the amount of air for the same effect can be reduced to 2·5% of that theoretically necessary. This is the basis of the possibility of replacing air fed directly into the melting chamber by super-heated steam, whereby the effect on furnace aerodynamics is similar, while a certain equalization of temperature which is obtained along the melting chamber (on account of the dissociation of steam and subsequent burning of hydrogen) in the second half of the heat, produces a positive effect. When compressed air or steam is fed directly into the melting chamber, the jets of air or steam must not disturb the organization of the main gas stream. Air and steam must, therefore, be fed in jets in the walls of the gas port and directed underneath the flame as low as is possible without choking up the downtakes with slag.

As regards the use of oxygen, the best method of feeding it into the flame is also by means of lances from the side of the gas port and directed underneath the flame (Fig.5). However, the influence of an oxygen jet is not limited by aerodynamic phenomena but is more complex since the field of concentrations substantially changes.

At present there are no OH furnaces with Moll and Maerz ports in the USSR but Soviet plants have carried out research on models. The simplicity of design of Moll ports does not compensate for their main defect, i.e. sensitivity to the ratio of air to gas velocities and their irregularity. Although the mixing of gas and air in Moll ports is very intensive, flame organization is bad. Twisted and curled flames with very poor bath-covering capacity are normal with

these ports. Maerz ports are also very simple in design. They can be used in conditions where flame organization in the melting chamber is ensured from a fuel point of view by the fuel stream (fuel oil, high-pressure natural gas). In the absence of such conditions, there is created, as it were, in the first half of the melting chamber, an independent movement of fuel and air streams, bad mixing conditions, and inadequate capacity of the furnace atmosphere for bath oxidation. The well developed recirculation underneath furnace roofs is absent with Maerz ports and this affects conditions under which the roof refractories operate. These conditions are more favourable in the first half of the melting chamber because of air cooling, and less favourable in the second half where an intensified transverse circulation causes overheating of the roof and intensifies the action of slag dust.

GENERAL CONCLUSIONS

Research into the aerodynamics of OH furnaces using models and also under production conditions has clarified very many questions which, in the past, have been very debatable.

It may now be asserted that, as a first approximation, a scientific basis of OH port design has been created. Further progress depends on the development of calculating techniques which are still far behind what is required in practice.

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The effect of flow distribution on air preheat in the open-hearth furnace

J. Chapman and W. Montgomery

SYNOPSIS

Investigations into the flow of air and waste gases through cold open-hearth furnace regenerators showed severe maldistribution. By the use of suitably disposed baffles and guides in the flues, regenerators, and on the bridgewall, it has been shown that improvements in distribution can be achieved. When these modifications were built into the furnaces, the air preheats were increased and savings in fuel effected. 1876

INTRODUCTION

WITH THE ASSUMPTION that increased air preheat temperatures favour an increase in the thermal efficiency of an OH furnace, an investigation was undertaken at the Templeborough melting shop of Steel, Peech and Tozer to determine existing preheat levels and the possibility of improvement.

Air preheat measurements had been taken at Templeborough in the past during special trials,[1,2] but no comprehensive survey had been attempted before in the air preheat and checker design field. Besides finding the normal level of preheat it was hoped to find the effect on preheat of several factors such as flow distribution through the checkers, air infiltration, not only in relation to volume but also the position of infiltration, and the adequacy of checker volume.

Furnace design

The furnaces are of 110-tons capacity. They have an arched roof, flat from end to end, and a single uptake 4 ft × 10 ft opening into a single slag pocket. These design features are common. It is from the bridgewall to the stack that the major differences occur. Eight furnaces have single checkers and flues (Figs.1 and 2), four furnaces have double checkers and flues (Figs.3 and 4), one furnace has double checkers with a single flue, and one furnace has a single checker with double flues.

Combustion control

The oil-fired furnaces are supplied from a ring main, where oil temperature and pressure are maintained at constant levels. The oil flow to individual furnaces is manually controlled. Maximum oil flow is 425 gal/h. This is carried until the roof temperature reaches 1650°C; the flow is then reduced to maintain this temperature or whatever level the type of steel being made demands.

A constant amount of air is supplied to each furnace. In the case of the single checker/flue furnace, air supply is by natural draught, and for the double system air is supplied by forced draught fans. The volume per hour is set from results of waste-gas analysis in the downtake during the period of full oil flow. Regulation of air is based on a 2–5% oxygen content in the waste gases. Frequent checks are made throughout each campaign to ensure that this range is maintained. Furnace pressure is controlled manually between the limits 0·04–0·06 in.wg.

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METHOD OF PREHEAT MEASUREMENT AND PRESENTATION OF RESULTS

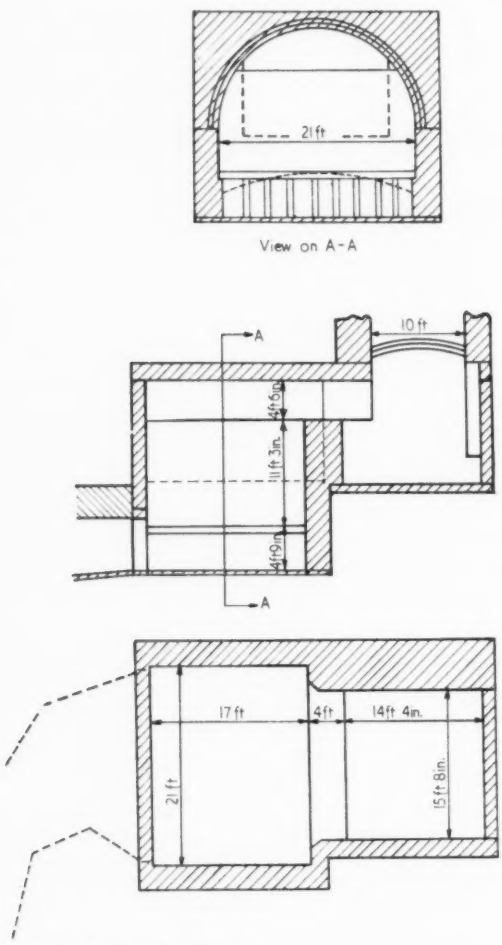
The suction pyrometer was placed through the pit side uptake wall 1 ft 6 in. above stage level and projected into the uptake 3 ft 6 in. from the side wall and 1 ft 6 in. from the end wall. All the temperatures were measured at one end of the furnace. An on–off valve in the compressed air line to the waste-gas ejector activated by the furnace reversal system permitted a 24-h record to be obtained with little supervision. Pyrometer hoods were changed about every 12 h or when necessary because of make-up. Very little trouble was experienced from this cause. If any fault or doubt occurred during any one cast, the whole record for that cast was rejected.

Air preheat temperatures were obtained using a BISRA-type internal-hood suction pyrometer as described elsewhere.[2] Temperatures were recorded on a Honeywell-Brown single point recorder.

Before this survey began, work had been carried out to establish the best position of the pyrometer for taking air preheat temperatures. It had been found that there was a temperature gradient up to about 9 in. into the uptake from the side and end walls. Beyond this it appeared that the temperature variation was so small it could be ignored.

To obtain the scatter bands shown in Figs.5, 6, and 8, mid reversal temperatures were used. This method was adopted rather than average 'area under curve' temperatures because it ruled out any uncertainty regarding the beginning of the reversal, and what may or may not have been the tail of the reversed waste gas.

Mid reversal temperatures have their disadvantage, however, in that they do not enable differentiation



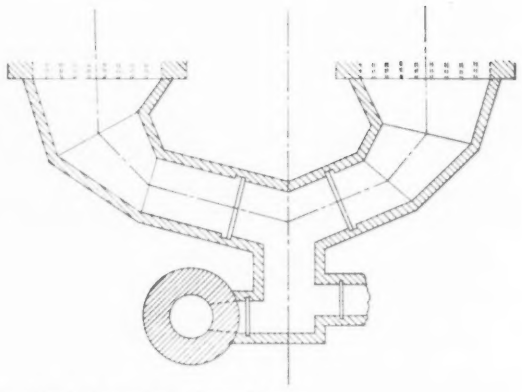
1 *Single checker and slag pocket*

between a curve with a high initial temperature and a fairly rapid decrease, or a curve with a lower initial temperature and slow decrease, both of which could give the same mid temperature and yet have different total values.

At least four complete casts were used to construct the scatter bands; this was regarded as the minimum number required to show variation after any alteration. Usually the four casts were spread over a 6–8 cast interval some time after the third or fourth week of furnace life.

Although the preheat was only measured at one end of the furnace, a check on the checker top temperature records showed if any serious unbalance occurred during the survey. If this did occur, the survey was halted until a reasonable balance had once again been achieved.

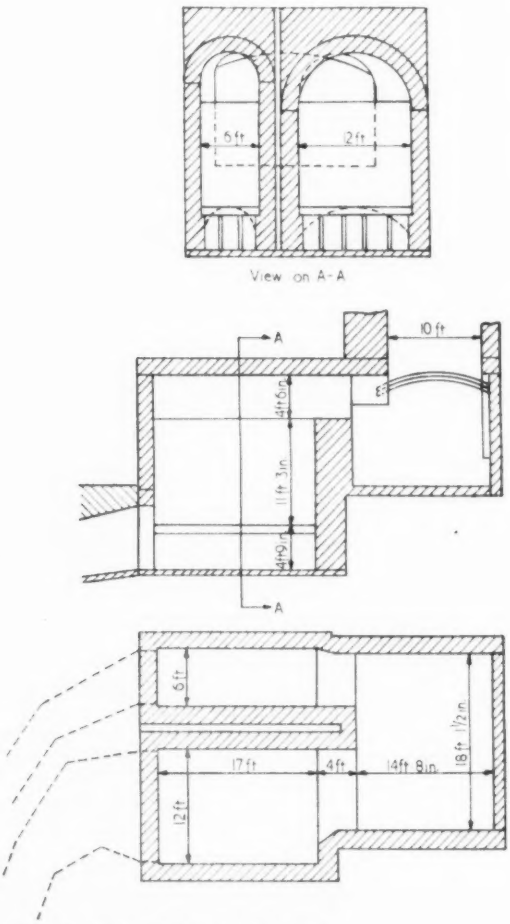
As will be seen in the section on 'Effect of furnace age', the air preheat generally appears to increase from the beginning of a campaign, until about half life, and then decrease towards the end of the campaign. The difference between each end, and mid life, of the campaign is about 60°C. This indicates that when making comparisons between campaigns it is essential



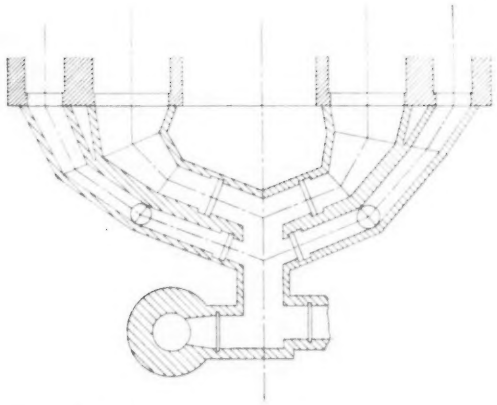
2 *Single flue system. Normal design*

that the respective measurements should be taken at about the same stage in the furnace life. Otherwise an apparent increase or decrease could well be due solely to the age of the furnace, and not to any alteration in design or operation.

When making comparisons after alteration, either



3 *Double checker and single slag pocket*



4 *Double flue system. Normal design*

sealing or constructional, it was considered unwise to attribute any difference in preheat to the alterations unless a consistent 20°C difference was obtained.

General level of air preheat

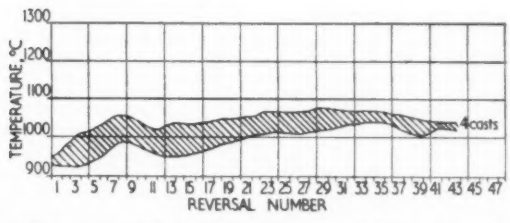
The first step in the investigation was to obtain some measure of the general level of air preheat and from this decide if any improvement were possible. The items considered as being profitable avenues of investigation were primarily air infiltration, flow distribution in the flues, checkers, slag pockets, and increased surface area of the checker filling.

Figure 5 shows results obtained on *S* furnace. This is a typical example and shows maximum mid reversal temperatures of about 1075°C. The main impression from these records was that the general level was disappointingly low.

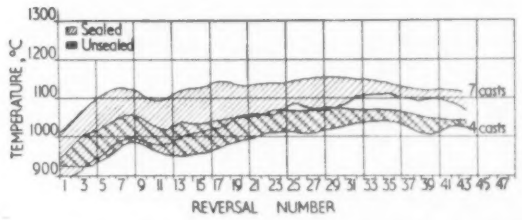
Air infiltration

Several investigators have emphasized the importance of the effect of air inleakage on furnace performance and fuel consumption and pointed out that attention to cracks in the brickwork, leakage round the damper frames and doors, is usually worth the effort. The first step towards increasing the preheat was a thorough examination of the 'downstairs' portion of the system and the sealing of any holes and cracks that showed air leakage when tested with a small flame. It was found that the checker and slag-pocket wickets were in need of attention, especially at the top of the wicket wall, but the largest infiltration was found round a cooling gap surrounding the slag pocket.

The slag pockets at Templeborough were built so that a 3-in. cooling gap existed between the side walls



5 *'S' furnace. Mid reversal temperatures. Preheat scatter band (unsealed)*



6 *'S' furnace. Mid reversal temperatures. Preheat scatter bands*

and floor of the chamber proper, and the slag collection chamber. The reason for this was to facilitate removal of the block of slag at the end of a furnace campaign and so avoid excessive damage to the permanent brickwork.

The air cooling gap was built so that the slag chamber was sealed off from the outside atmosphere but the gap itself open to the atmosphere, allowing a circulation of air around the slag block. After the furnace had been in operation a few days, cracks appeared in the seal, and fairly large quantities of air were drawn into the slag pocket through the cooling gap.

Large differences in air infiltration on the waste-gas cycle, determined from the change in waste-gas analysis measured simultaneously at the uptake and waste gas culvert, were found. The reason for this was ascertained when subsequent examination of several slag pockets revealed that portions of the false wall had collapsed on some furnaces.

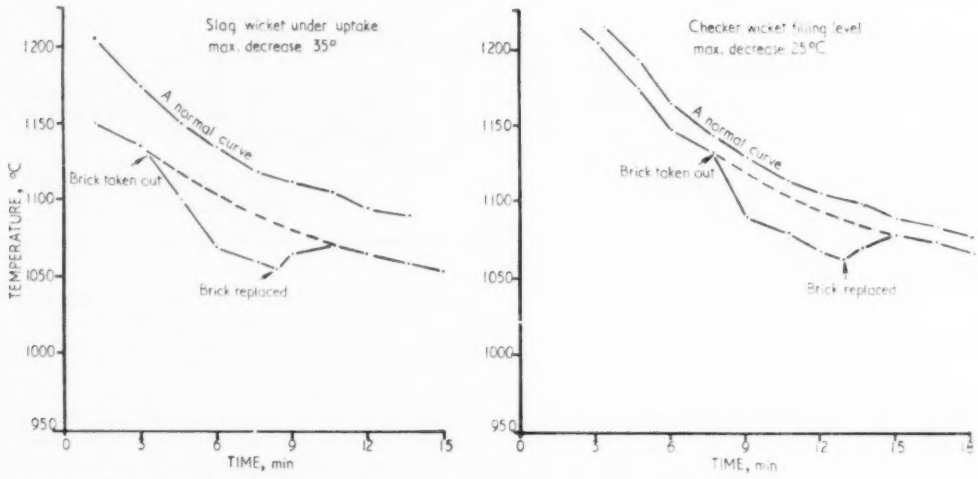
After measuring the air infiltration and associated air preheats on several furnaces, the air gap round the slag pocket was sealed off from the atmosphere, slag and checker wickets sealed, and the wicket faces coated with clay. These seals were left to dry and settle for two or three days before the infiltration and preheats were measured again. The effect of this sealing is illustrated by Fig.6. It can be seen that the preheat increased by about 80°C. This was associated with a reduction of infiltration of the order of 80000 ft³/h.

To determine the effect of volume and position of inleakage a few simple experiments were carried out on a sealed chamber and slag pocket. The experiments consisted of opening a hole of known area in different parts of the system, and measuring the immediate effect on the preheat. At the same time the pressure difference was measured, and from this the amount of air inleakage calculated. The results are shown in Table I and the temperature curves in Fig.7.

The temperature drop due to the hole in the slag pocket is higher, presumably due to lack of mixing in

TABLE I Deliberate infiltration

Position	Pressure diff., in. wg	Hole size, in²	Volume in leaking, ft³/h	Immediate effect on preheat, °C
a Slag pocket wicket	0·20–0·25	72	21000	−35
b Checker wicket top of filling	0·20–0·25	72	21000	−25
c Damper frame		40		Nil



7 *Deliberate infiltration. Immediate effect in preheat*

the short vertical distance from the hole to the pyrometer (12 ft). For the checker wicket hole the temperature drop is about equal to the calculated drop for the conditions existing at the time, assuming complete mixing. The shortest combined horizontal and vertical distance from the checker wicket hole to the pyrometer was 40 ft.

One point of interest arose during the sealing trials regarding the importance of sealing all the points of actual and possible leakage. When one furnace had been sealed, as described above, very little difference was found both in inleakage and preheat. This was surprising in view of the size of some of the gaps that were sealed. On further examination it was found that a much smaller wicket in the side wall of the slag pocket had been neglected. When this smaller wicket was sealed, the infiltration fell by 90000 ft³/h and the preheat increased by about 70°C.

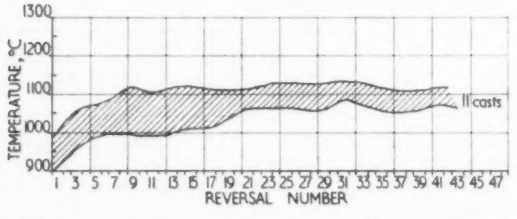
Double checkers v. single checkers

The scatter band shown in Fig.8 was obtained from *L* furnace. This furnace, operated at the standard oil flow of 425 gal/h, has retained the old gas and air checkers and double flues, but has a single slag pocket and single uptake. On comparison it will be seen that the preheat obtained from this furnace is at least as good as, and occasionally better than, the preheat obtained from the single checker furnaces, this in spite of the total checker volume being 16% greater in the single checkers than in the two checkers.

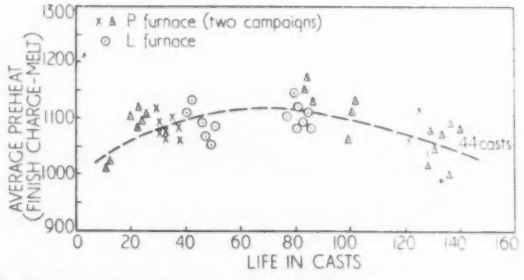
Three contributory factors are thought to cause this result. Although the total checker filling volume is 16% greater in the single checker, the plan area at checker top level is 3% less than that provided by the double checker. This is due to the semi-arch of the checker chamber roof in the single checker creating a 'shadow' effect at the edges of the chamber (Fig.1). Another factor aiding the attainment of a high preheat in the double-checker double-flue design is the better overall distribution of air passing through the checkers. The third factor causing the levels of preheat to be similar in double checker and single checker furnaces is considered to be the path taken by the incoming air after it leaves the checker chamber. In the single checker the bulk of the air appears to flow straight into the pitside of the uptake, having little contact with the slag pocket walls. This is due to the highest point of the semi-arch, or the largest area over the bridgewall and the entrance to the uptake, lying in the same vertical plane, so that maximum flow out of the checker chamber occurs almost under the uptake. In the double checker chamber design, the uptake is in roughly the same vertical plane as the dividing wall between chambers. Thus neither of the two smaller air streams is able to flow straight into the uptake, but must undergo some mixing in the slag pocket before entering the uptake. Because of this, greater contact with the slag-pocket walls is inevitable, with the result that the air preheat is improved.

Effect of furnace age

The deterioration of output and economy, that generally occurs with the ageing of the furnace, can be



8 *'L' furnace. Mid reversal temperatures. Preheat scatter band*



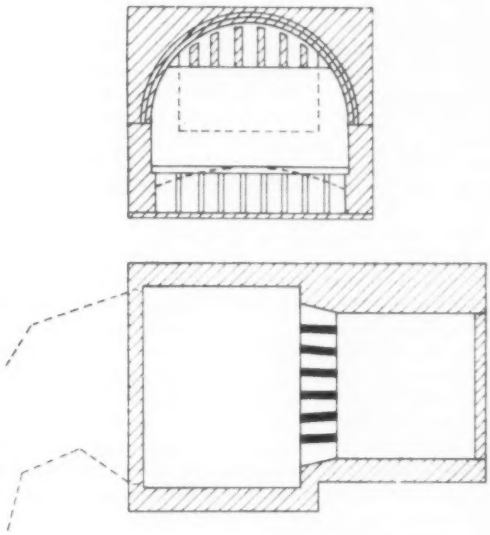
9 *Effect of furnace age on preheat level*

TABLE II First partition trials, 'F' furnace

Number	Outer culvert, ft³/h	Inner culvert, ft³/h	Total flow, ft³/h
1	300000	410000	710000
2	390000	390000	780000
3	500000	260000	760000
4	260000	500000	760000

attributed to factors such as checker blockage, air infiltration, eroded walls and roof. Most of these factors are difficult to measure during furnace life, and the proportion of the individual effects to the whole is difficult to determine. However, the collective effect can be said to be reflected in the air preheat level and from a record of this, throughout the furnace life, the relative efficiency may be determined. Figure 9 shows the average air preheat from finished charging to melted, plotted against furnace life in casts. It can be seen that the preheat level rises steadily to about the 80th cast, roughly halfway through the life of a Templeborough furnace, and then falls steadily to the end of the campaign. This trend is in close agreement with that reported by Carney *et al.*[3] They attributed the trend partly to the change in the air infiltration during the life. That is, as the inside brick surfaces of the walls become glazed, so the infiltration is reduced, then, as the refractory becomes thin and worn, new cracks appear and the infiltration increases again.

From a study of the deposits in the checker fillings at the end of the furnace life, and a knowledge of the flow distribution in the checker, discussed in a later section, the checker blockage may not have the adverse effect that the amount of deposit might suggest. From the geometry of a checker system it is obvious that the combustion air will tend to flow up the bridgewall side, and the waste gases down the flue side of the matrix. Any make-up of the checker during the campaign will tend to force the paths of these two streams closer together. In these circumstances the efficiency of regeneration in that part of the filling in use will be quite high, probably with an overall efficiency as high as a newly built checker, considering the very poor distribution in a new checker. From this, excluding reduced draught caused by the blockage, the reduction in preheat due to blockage may well be quite small and the major cause of reduction attributed to air inleakage.

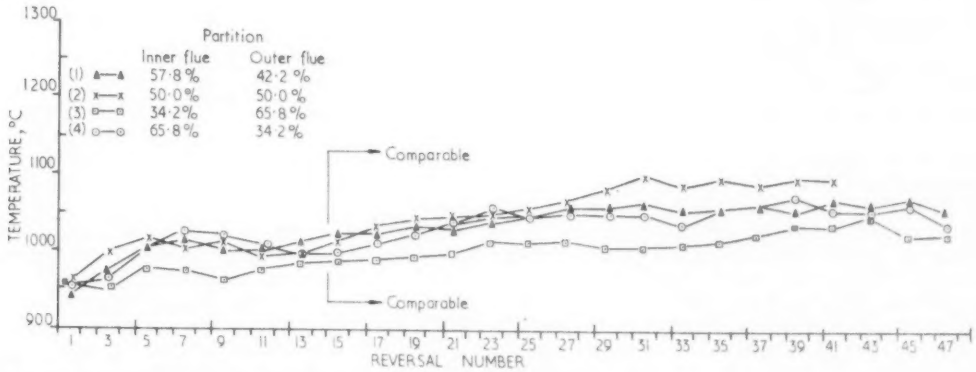


11 *Flow deflectors (mark 1) on bridgewall of 'P' furnace*

FLOW DISTRIBUTION

Partition trials

Velocity traverses in a cold newly built furnace, and water model work at Swinden Laboratories very quickly showed poor distribution. One aspect of the distribution in need of attention was that of the tendency of the incoming air to concentrate in the outside bend of the flue, creating a poor distribution in the checkers. Following this, further work on the water model showed that positioning a deflecting wall in the flue considerably improved the distribution. As described earlier, two of the furnaces at Templeborough are of a hybrid design. One of them, *F* furnace, has the original separate gas and air flues but with a single checker chamber. This condition proved to be helpful because it was immediately possible to test the water model findings, on the plant, without waiting for a furnace reconstruction to include deflecting walls.



10 *'F' furnace. Air partition trial on double flue, single checker furnace*

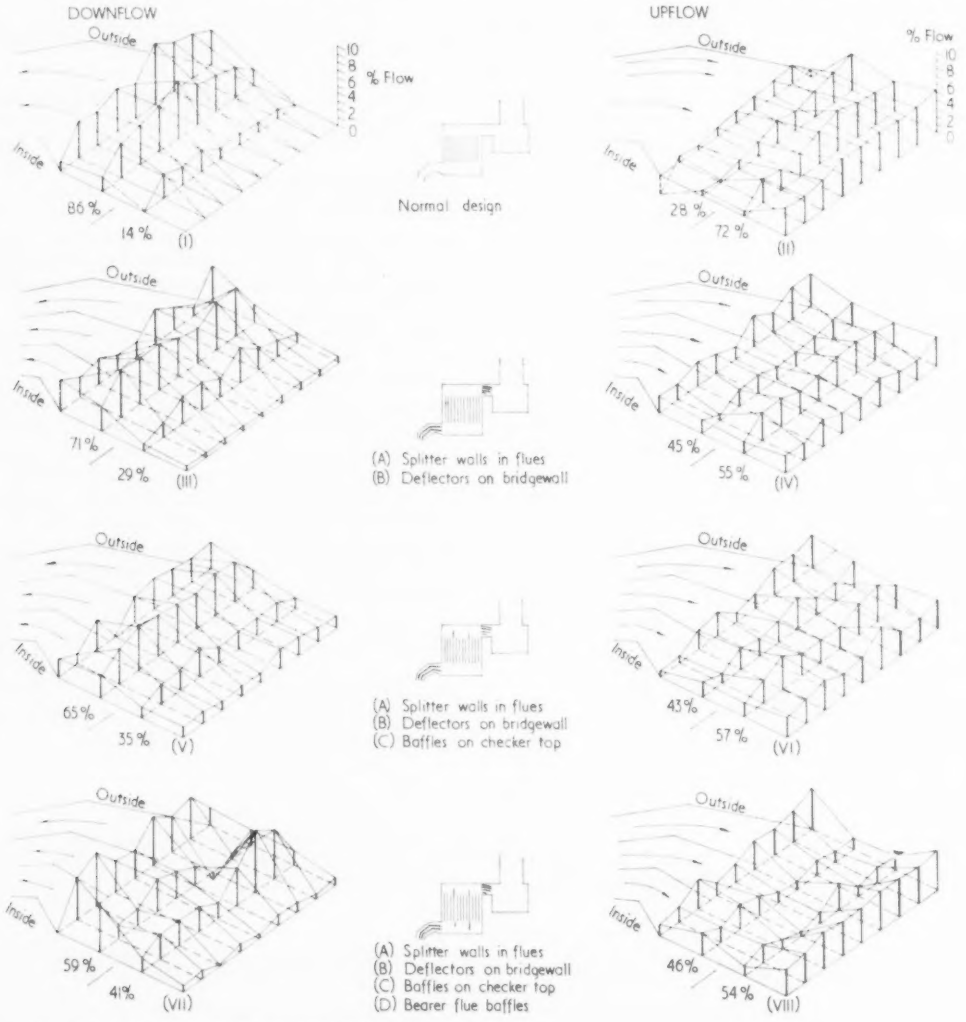
Each flue in *F* furnace has a separate air fan, thus making it possible to control the amount of air in each flue. Because of this a large proportion of the total air requirement could be blown into the outer or 'gas' flue thereby reproducing the conditions in the single flue design. Other proportions could be used at will to determine the best partition when coupled with whatever waste gas distribution existed on *F* furnace. This waste gas distribution was constant and uncontrollable.

Four conditions of air flow were investigated, as shown in Table II. For each trial, before the air preheat was measured, waste-gas analyses were taken and any adjustments to the air flow made to ensure that combustion conditions were about the same in all cases. Figure 10 shows the average values from at least four casts in each condition. It can be seen that little difference occurred between conditions 1, 2, and 4, the best condition being the one with equal flows through each culvert as in 2. The worst condition was 3, when 500000 ft³/h of air was passed through the outer culvert. This condition was thought to have the nearest similarity to the flow in the single flue design. This experiment was repeated on the next build, when the furnace was new, and the effects were similar. It is worth noting that during the rebuild the slag pockets were sealed. Air infiltration was reduced, thus higher volumes of air were indicated as passing through the valves and checkers to give similar waste-gas analyses.

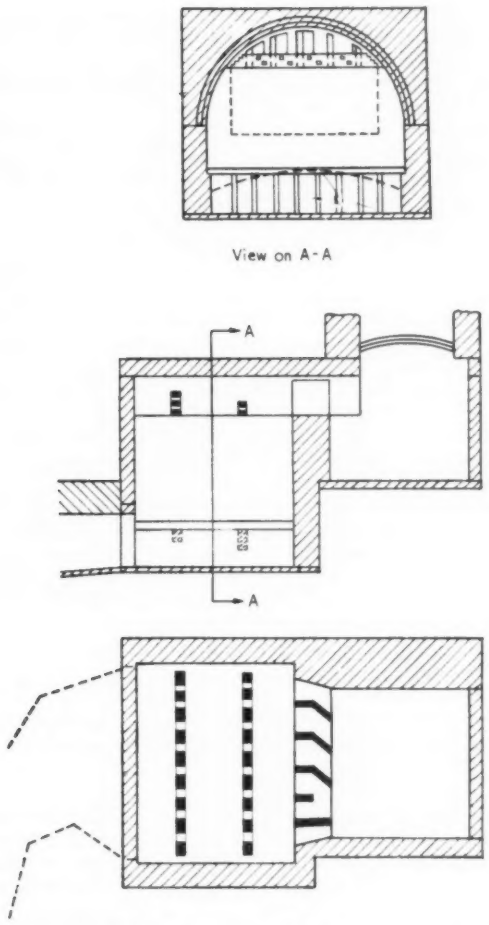
Apart from the fact that these experiments indicated the best way to partition the air in *F* furnace, they emphasized the results obtained at Swindon Laboratories and indicated the need for further experiment and development in the flow distribution field.

Deflector walls

Following a report by Goldenburg[4] on the use of flow distributors on the bridgewall, it was decided to use these in a modified form on *P* furnace. Goldenburg's intention had been to deflect the gases into a part of the chamber that was otherwise starved, due to a sudden enlargement when passing from the slag pocket to the checker chamber. The arrangement on *P* furnace



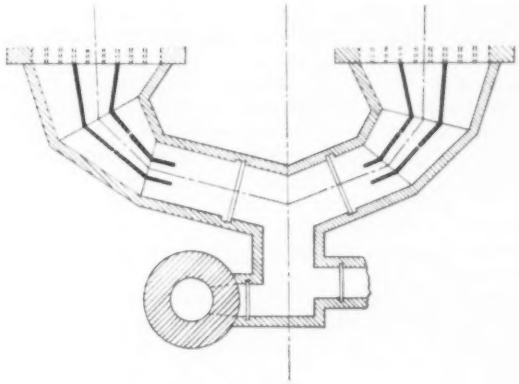
12 *Flow distribution in single checker system. TMS furnace*



13 *Bridgewall deflectors, checker baffles, and bearer flue baffles, in single checker system*

shown in Fig.11 was intended to break up what was considered to be an undesirable flow into the checkers from the slag pockets. The flow from the slag pocket into the checker chamber takes the form of twin vortices, the result of flow from a single uptake positioned roughly in the centre of the slag pocket when the slag pocket and checker chamber centre lines are common. Visual observations on the furnace showed that the vortices were destroyed by the deflectors, but distribution surveys showed that little effect was made on the waste gas distribution.

Although the deflectors had a negligible effect on the distribution in the checkers, the level of preheat obtained proved to be the highest measured up to that time. A maximum average of 1160°C was recorded. Unfortunately for comparative purposes, no record had been obtained on *P* furnace before the deflectors were built, so that only a limited sort of comparison was available, that with other furnaces. Because of the lack of data on *P* furnace, and the results of the distribution tests, it can only be assumed that whatever improvement did take place was due mainly to the extra heating surface provided by the deflectors themselves, plus some small amount from the slightly



14 *Splitter walls in single flue system*

improved distribution. This was substantiated by calculation from the report of the BISRA Regenerator Sub-Committee.[5]

As a consequence of the disappointing results from the distribution tests and water models following the installation, the *P* furnace design of deflectors was scrapped. From the results of further distribution tests on cold, newly built furnaces, using a technique described by Hulse,[6] a new series of baffles and guides was designed and tested.

Baffles and guides in single systems

Tests in a newly built cold furnace showed considerable unevenness in the distribution of waste gas and air into the checker filling. As well as a poor distribution individually, the waste gas and air were found to be badly distributed with respect to each other; diagrams, (i) and (ii) in Fig.12 show this quite clearly. Following these tests, a further and more extensive survey was made to determine the effect of various baffles and guides on the distribution of waste gas and air in the regenerators and flue systems of the single and double checker furnaces. Cold air was used as the test fluid to represent incoming air and outgoing waste gases.

Method of testing

The method of testing the individual, and collective, effects of the features shown in Figs.13 and 14, was to set the air or 'waste gas' flow to the desired value by manipulation of the waste heat boiler fan and the relevant dampers, thereby simulating natural air as used in the single system. In the double systems the forced air fans were used for the incoming tests, and the boiler fan used for the outgoing tests.

A 'Velometer', giving an instantaneous direct reading of air velocity, was used for the distribution survey. The air velocity was measured at four sections, i.e. entrance, $\frac{1}{4}$, $\frac{1}{2}$, and $\frac{3}{4}$ distances, in each of the nine bearer flues under the checkers. In the plane of each section nine readings were taken to obtain the average velocity at that section. In all, 324 readings were required to complete one survey, assuming repeats were not required, but because of some instability in the air flow this was never the case, and as many as 700 readings were normally required for a single

survey. The volume flow in any section was determined by multiplying the average velocity by the cross-sectional area of the bearer flue. Then, by difference, the volume up or down into the plan area between the sections could be determined. These volumes were determined as a percentage of the total flow. The results are presented in the diagrams in Fig.12.

The air volumes used were calculated to give approximately Reynolds similarity at the bridgewall section for waste gas flow, and the flue fantail for incoming air.

Owing to the temperature variation across a checker and the changes that take place during a reversal, it is not possible to conduct a cold trial under true similarity conditions for all parts of the system. For instance the flows required for true similarity vary from top to bottom of the filling as 1·3:1.

Testing a range of flows each side of the calculated value showed relatively little change in flow distribution. The calculated flow rate for exact similarity was 400000–500000 ft^3/h, but to obtain more accuracy in regions of low velocity the majority of tests were carried out using flows of 750000–800000 ft^3/h.

Four features were tested, as shown in Figs.13 and 14, comprising: (*a*) splitter walls in the flues, (*b*) deflectors on the bridgewall, (*c*) baffles on top of the checker filling, (*d*) baffles in the bearer flues.

Splitter walls in the flues It is well known that fluids flowing round a bend tend to concentrate on the outside of the curve. This is precisely what happens in the TMS flues, causing an asymmetrical distribution at the base of the checkers (*see* (i) Fig.12). The splitter walls were intended to improve the transverse distribution in the flue, before the incoming air entered the checker chamber.

As mentioned earlier, work on water models at Swinden Laboratories, and experience on *F* furnace, had shown that splitter walls in the flues would provide some improvement in the flow distribution of the incoming air and an improvement in the preheat. For the water model trials a short-angled wall had been positioned in the approaches to the bearer flues and its position had been found to be rather critical. In the trial furnace two splitter walls were used to divide the main flue exactly into three, and were taken back towards the chimney almost as far as the valve arch, to ensure collection of equal, or near equal, amounts of air from a section where it was known that the distribution was fairly even, i.e. before the first bend.

This method of collecting equal amounts in each division only applies using natural air, when the air is drawn in all round the lid of the air box, thus providing a fairly constant transverse distribution in the first leg of the flue. When using the forced draught fan, maldistribution is caused at the air box, due to the position of the inlet at one side of the box. The momentum of the air carries the stream across to the opposite side of the box, from where it flows into the flue providing a large flow along the inside wall with very little at the outside wall.

Owing to the shape of the flue, this condition is reversed at the first bend, where the air stream flows across to the outside wall. From this point there is little difference in transverse distribution between natural and forced air.

It can be seen from Fig.14 that if forced air were to be used, the splitter walls as shown would merely reverse the conditions existing before they were built; most of the flow would then be confined to the inside wall. This may be avoided by determining the boundaries of the main stream and progressively shortening the splitter walls away from the valve until each one projects far enough into the stream to collect one-third of the total flow into each passage.

Flow deflectors on the bridgewall These deflectors were designed to provide an even transverse distribution of the outgoing waste gases. The angled portion was intended to direct the incoming air round the walls of the slag pocket before it entered the uptake.

It has been shown from water model work that, in the normal design of a single checker chamber, single slag pocket furnace, the incoming air tends to flow from the checker chamber into the uptake, without the majority of the flow coming into contact with the slag pocket walls. As suggested in an earlier section, this results in a comparative low preheat and detracts from the value of the larger volume single chamber.

A further feature of the flow in the normal single checker design is that of a pitside bias in the uptake, due to the method of entry into the uptake. This causes asymmetry in the furnace chamber resulting in an air rich flow on the front wall, and a fuel rich flow on the back wall. It is reasoned that this fuel rich back-wall flow greatly aggravates the roof guttering experienced along the skewback, and over the taphole. In an attempt to counter the asymemtrical flow in the furnace chamber, the burner is normally set off-centre towards the front wall, by some 9 in.

As well as improving the preheat by causing greater contact with hot surfaces in the slag pocket, the deflectors were expected to provide secondary benefits by removing the uptake bias, such as less guttering of the roof over the back wall. That this does appear to be so is supported by the fact that, on the single checker furnaces, the amount of roof repair decreased considerably when the deflectors were built. Of the furnaces so far included in the trials, *M* furnace had no roof repairs at all during the two campaigns, with a small repair at the 13th week during the third campaign. Similar improvements have occurred on other furnaces, in that any roof repairs have been smaller than normal, and have occurred later in the campaign, i.e. at the 11th to 13th week rather than the 8th to 9th week, as on furnaces without the deflectors.

Baffles on top of the checker filling These features were intended to improve the longitudinal distribution of waste gases shown in Fig.12(i).

It was thought that the main effect of these baffles would be a deflection from the front of the wall, down into the filling. To avoid recirculation immediately behind the walls, and risk of consequent deposit, holes were provided at regular intervals along the length. The walls were also racked in at the edges to compensate for the shadow effect of the semi-arch. Experience has shown that the recirculation that does occur behind the walls, in spite of the holes, is a major contributor to the improvement provided.

For trial purposes mock walls of hardboard were used, mainly to facilitate positional variation. With these baffles it was only a few minutes work to make a

complete re-arrangement at any position on the filling top. By having a unit construction, any variation in setting could be tested.

Baffles in the bearer flues Because the waste gas and air streams normally utilize opposite halves of the filling, any adjustment to the distribution of waste gases on top of the filling should be accompanied by a similar adjustment to the incoming air in the bearer flues. The splitter walls in the main flue provided an effective transverse distribution, and the bearer flue baffles were intended to correct the longitudinal distribution by deflection, in a similar manner to that expected of the top baffles.

Although the bearer flue baffles were tested in their turn, it was known at the outset that the melting shop management were uneasy about their inclusion as a permanent feature. This was justified because no information was available, or obtainable, regarding the effect the baffles might have on the oxide deposit that occurs in the bearer flues. In the normal design the largest deposits occur at the exit end of the bearer flues, due to the expansion into the main flue, plus a sudden change of direction when turning out of the filling. In this position the pile of deposit could become quite large, and yet the furnace could remain workable.

Discussion of results

The results are depicted in isometric form in Fig.12 (i–viii), the base of the diagrams representing the checker plan area, and the vertical ordinates representing the volumes between the measuring sections. The sections are numbered 1–4: no.1 is the wicket section, no.4 the bridgewall section.

Normal checker: waste gas flow (*i*)

Figure 12(i) shows quite clearly the marked asymmetry of the outgoing waste gas. The largest concentration of flow occurs at the wicket end of the chamber, in sections 1 and 2.

The area between sections 3 and 4 is noteworthy for the considerable reverse flow in the whole of the section. This can be attributed to the high velocity with which the waste gases flow over the bridge wall. No opportunity is provided for recirculated gases behind the bridgewall as occurs with the checker top baffles, because the filling is level with the bridgewall. In these circumstances the recirculation consists of drawing gases up into sections 3 and 4 from the portion of gases that has already reached the bearer flue, by way of sections 1 and 2.

If the checker be divided vertically into two parts so that sections 3 and 4 comprise one part (bridgewall), and sections 1 and 2 the other part (wicket end), then the division of gas flow into these parts is 86% into the wicket end half with 14% at the bridgewall end half.

The effect of the semi arch may be discerned in this and other downflow diagrams, in that the flow down the outside and inside walls is less than in the centre regions.

Normal checkers: in coming air flow (*ii*)

The ingoing air flow distribution is practically the reverse of the downflow conditions. The highest flows occur in the bridgewall sections 3 and 4, with a further bias to the outside of the chamber, due to the effect of the curved main flue.

The ingoing air division of flow into the two parts is 28% wicket end, and 72% bridgewall end. Thus 72% of the incoming air has available, in the filling, only 14% of the heat from the waste gases. This situation is in very close agreement with that reported by Marsh.

APPLICATION

Application of splitter walls in the flue (*A*) *and deflectors on the bridgewall* (*B*)*: waste gas flow* (***iii***)

The most important effect of these features, mainly the deflectors on the bridgewall, is that the reverse flow in the bridgewall section has been eliminated and a small positive flow provided.

It may be observed that if the distribution is dependent on velocity over the bridgewall, and the presence of deflectors increases that velocity by reducing the area, then the distribution should in fact be worse.

The answer once again is in the provision of recirculation areas, this time immediately behind the deflectors and since the new design contains a fairly steep angle the recirculation is much stronger than that provided in *P* furnace where only a very small angle existed.

This recirculation area has been shown to exist both on the water model and by velometer studies on top of the checker filling.

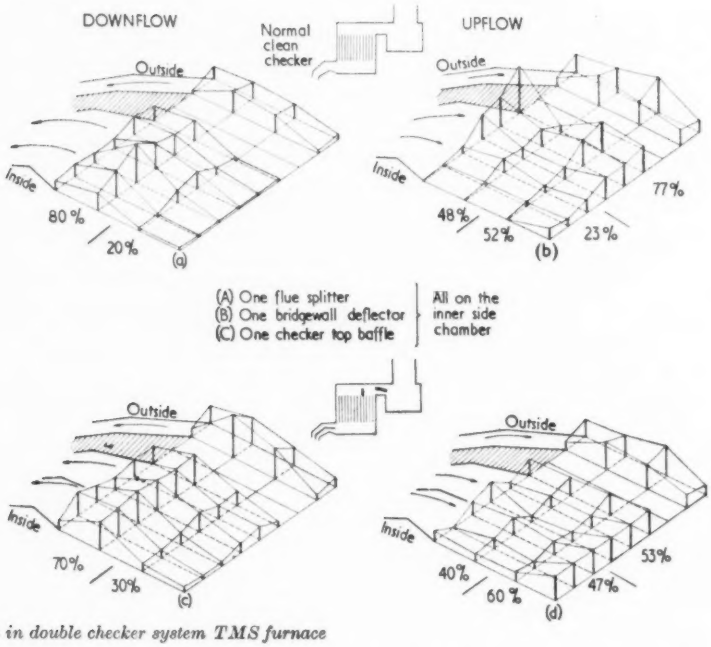
As well as elimination of the negative zone at the bridgewall, the high flow at the wicket shows a considerable reduction, with section 2 showing some increase. This is solely due to the increased velocity caused by the reduction of area over the bridgewall. The gases strike the wicket and flow back over the filling towards the bridgewall, forming a large unstable vortex, roughly between sections 3 and 2, and flow down in these sections. The instability of the vortex is mainly in a transverse plane. This may be seen by the two peaks in sections 2 and 3 and was certainly detected in the velometer readings during the survey.

The ratio of flow in the two vertical parts has improved from 86/14 to 71% in the wicket half, and 29% in the bridgewall half. That the transverse flow improved considerably was shown by the proportions 25, 26, and 22 flowing into the three splitter flues.

A + *B*: *incoming air flow* (*iv*)

The most obvious improvement is the equalization of the transverse distribution in the chamber, due to the splitter walls. The effect of the bends in the splitter walls can be seen in the minor peaks on the outside line of each wall; this continues to a lesser degree through to the bridgewall. The lesser peaks are only to be expected, for what has been achieved is to change one large asymmetrical flow into three smaller asymmetrical flows. The air volumes flowing into each splitter section are not quite equal. From outside to inside the proportions are 34, 28, and 25.

An unexpected improvement has occurred in the longitudinal distribution compared with the normal checker. This is thought to be due to a disturbance in the flow pattern, caused by the bridgewall deflectors, and was first noticed on the *P* furnace trial when only the bridgewall deflectors were used. A similar effect is



15 *Flow distribution in double checker system TMS furnace*

obtained by reducing the flue size of the checker filling, but this is accompanied by an increased tendency to blockage. One purpose of this paper is to show that additional constructions can improve the checker distribution, without the danger of increased blockage. Here the distribution ratios are 45% and 55% wicket to bridgewall.

Application of A and B plus checker top baffles (C): waste gas flow (v)
Addition of the checker top baffles has further increased the flow down the section adjacent to the bridgewall. So that now the ratios are 65% down the wicket half, and 35% down the bridgewall half.

In spite of the improvement obtained so far, i.e. from 86/14 to 65/35, the effect of the checker top baffles was disappointing. Sections 3 and 4 show some improvement and a levelling off has occurred in sections 1 and 2 but the flow here is still rather high.

A+B+C: incoming air flow (vi)
This condition is considered sufficiently like A+B that for practical purposes they are the same. The proportions are 43% and 57%, compared with 45% and 55% for A+B. This difference probably represents the margin of error that can occur under the conditions of test, and with the instrument used. From the similarity with A+B it may be inferred that the top baffles have little effect on the incoming flow.

Application of A, B, and C plus baffles in the bearer flues (D): waste gas flow (vii)
Instead of continuing the progressive improvement obtained in the other two cases (A+B, and A+B+C) the addition of the baffle walls under the checkers caused the reversion shown in Fig.12 (vii). One possible explanation of this is that C and D in some way interfere with each other and to a large extent cancel out, providing a condition not unlike that of A+B. A further possible reason for this result is that the bearer baffles create local recirculation in the three pockets formed by the two sets of baffles. In these circumstances this recirculation would be difficult to detect and allow for.

A+B+C+D: incoming air (viii)
Here again the expected progressive improvement did not occur, although the reversion is not so serious as that of the downflow condition.

Peculiarly enough the proportion of flow into each half in this example and in the downflow condition provide the best division so far obtained; 46/54 for incoming air and 59/41 for the waste gases. Even so, the diagrams show that these proportions are ill-matched if smaller sections are selected for comparison.

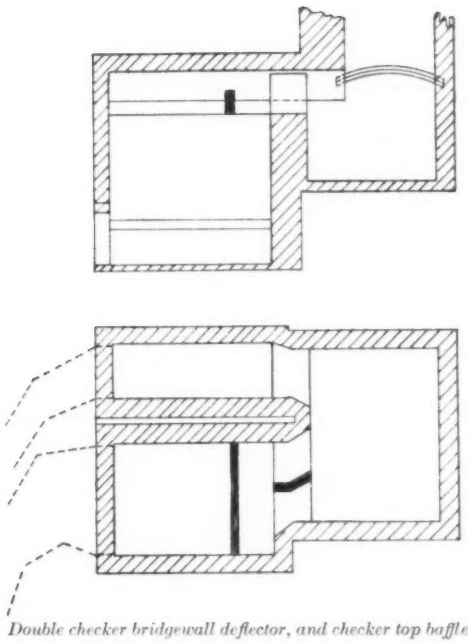
Choice of features
From all aspects the downflow of waste gases provided by A+B+C is considered to offer the greatest improvement. While from the upflows that provided by A+B is the better.

However, as the conditions must be considered in pairs, those of A+B+C provide the highest total improvement and those of A+B come second. The features A+B+C have been included in *M*, *S*, and *P* furnaces and A+B in *N* furnace. In all other respects these furnaces are of the same design.

DISTRIBUTION IN DOUBLE CHECKER FURNACES

The oil-fired double-checker double-flue furnaces *H*, *K*, and *L* are all worked with forced-air fans with a separate fan for each flue. This is in contrast to the single system furnaces already described which are worked with natural air.

Because of the difference in air feed, the double checker distribution surveys were carried out using two methods for providing the test air; the boiler fan



16 *Double checker bridgewall deflector, and checker top baffle*

pulling air down through the fillings to simulate waste gases, and the forced air fan blowing air into the filling to simulate ingoing air.

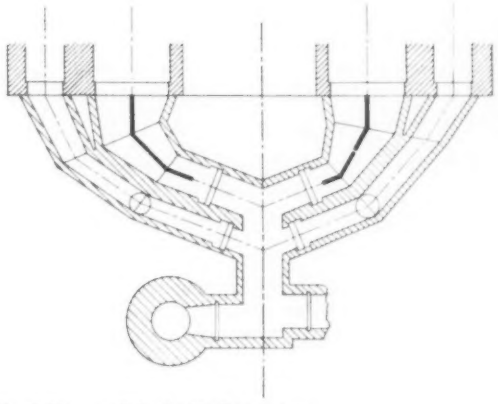
It was expected that the distribution in the larger checker of the double checker system would merely be a small scale edition of the single checker. Figure 15*a* shows that this is substantially so with one important difference. On the incoming cycle the same tendency for bridgewall end and outside wall bias is apparent, but the important difference lies in the volume of flow in the wicket section. It can be seen to be as great as that flowing up the bridgewall section. The reason for this difference was not apparent until it was noticed that the floor of the checker chamber was on a higher level than the floor of the main flue, the two being connected by a sloping ramp. The same is true for the single flue system but the ramp is much steeper in the double system.

Because of the method of entry from the fan into the flue, the incoming air tends to concentrate on the floor of the flue; in these circumstances the ramp has a much greater effect than if the incoming air had a more even vertical distribution. The incoming air behaves as though the ramp continued into the filling, and thus the bearer blocks become baffles of decreasing proportions along each bearer flue. This ramp, therefore, appears to be an answer to the problem of baffles in the bearer flue, in that it has no effect on deposits under the filling.

Mainly because of the size of the smaller checker and also the complex flow into it, no efforts were made to alter the distribution. In this smaller chamber, the position of the large volumes downwards is the same as the larger volumes upwards, so that though the filling is not used to its fullest, the parts that are in use are used efficiently.

Baffles used in double system

As described earlier, the double checker dividing wall



17 *Splitter wall in double flue system*

does some service as a bridgewall deflector; it prevents the incoming air from flowing straight into the uptake.

However, it was considered that the flow of waste gases into the larger chamber required some correction, as it tended to have a bias towards the inside wall and roof. Instead of a double vortex, as in the single chamber, the double chambers had one vortex into each, the vortices revolving in opposite directions. Mainly because of the chamber size, only one checker top baffle wall was used in the larger chamber, in conjunction with one bridgewall deflector.

Examination of Figs.15*b* and *c* shows that the transverse distribution has improved due to the deflector, but that the longitudinal improvement is not as much as experience on the single chamber would suggest. It was difficult to direct the flow, as most of the gases passed near the roof as they entered the chamber.

The improvement obtained in the incoming air distribution was better than that for the outgoing waste gas. In view of the effect of the flue ramp, addition of a single splitter wall was all that was necessary to change the pattern from that in Fig.15*b* to that in Fig.15*d*. The addition of the single deflector on the bridgewall had a much smaller effect in these circumstances than that found in the single system. To quote the distribution improvement in percentage ratios, it became necessary to divide the air checker two ways as shown in Fig.15. If the style for the single system is used, the ratios appear worse in Fig.15*d* than in *b*, but if a division is made in the longitudinal direction, the true improvement can be seen, i.e. from 23/77 to 47/53.

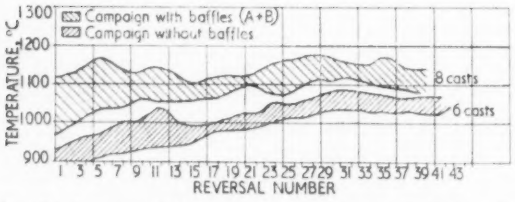
The baffle arrangement as shown in Figs.16 and 17 providing the distribution pattern in Fig.15, have been incorporated in *H*, *L*, and *K* furnaces.

FACTORS AFFECTING THE RESULTS IN PRACTICE

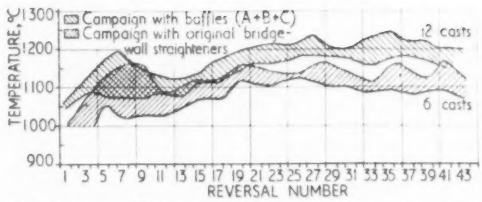
Deposit in filling and bearer flues

In general the deposits experienced in the Templeborough checkers would be expected to work in the opposite direction to buoyancy, in their effect on distribution.

On the outgoing cycle, the major portion of the deposit tends to be in the wicket section of the chamber, though this will be modified by the presence of baffles. As this deposit accumulates, increasing the pressure drop through that section of the checker filling, the outgoing flow will be forced more and more



18 *'N' furnace. Mid reversal temperatures. Preheat scatter bands*



20 *'P' furnace. Mid reversal temperatures. Preheat scatter bands*

to the centre and bridgewall sections, where resistance in the filling is less. The change of outgoing distribution during the life of the furnace is, therefore, in the desired direction.

The effect of deposit on the incoming air cycle will be a resultant of two features. One, the increased resistance of the checker filling at the wicket end, and two, the effect of a pile of deposit at the exit of the bearer flues. These two will increase in magnitude together, as the furnace ages, and will probably move the ingoing distribution away from the desired direction.

Buoyancy

When testing the various modifications using cold air as the test fluid, some allowance may have to be made for the effect of buoyancy in the working furnace. The effect of this is not precisely known.

PERFORMANCE AND RESULTS

Constructional

The first splitter walls were of 4½ in. firebrick, without any expansion allowance along the length. Because the walls were free at one end it was thought that they would expand down the flue. In fact the angle corners along the walls were so sharp that the expansion of each section pushed the walls outwards at the corners, thus resulting in part collapse. As well as expansion failure, the walls themselves proved to be too high for stability at 4½ in. thickness. More recently the splitter walls have been built 9 in. thick for the first 4 ft in height and 4½ in. above, with two expansion gaps along the length.

Both checker top baffles are 9 in. thick, buttressed on each side in four places. These baffle walls are built dry, of best quality silica, but with jointing at the buttress/wall junction, and where necessary to maintain a level. Holes are provided by leaving out every alternate header in the alternate courses.

The life of the baffles varies between 8 and 13 weeks for the smaller front baffle, with 10 to 13 weeks for the larger baffle. Failure appears to be mainly by attrition for the small wall and by collapse or part collapse for the large wall, due it is thought, to settling movement of the checker filling. Examination of the filling after a campaign has shown that blockage of the line of checker flues immediately in front of the walls is usual, with the odd slagged over pile of bricks partly blocking an area of four of five flues. Comparing the volume of brickwork in the new walls with the amount left after a campaign, the proportion of blocked filling is surprisingly small.

Bridgewall deflectors are 9 in. thick, of first-quality silica brick, with best silica jointing. Their life ranges from 9 to 13 weeks, the damage being mainly attrition and melting. In the earlier designs the deflectors were 9 in. thick throughout, until it was found that they collapsed prematurely because of undercutting at the slag pocket leading edge. This has now been remedied by including an 18 in. toe at the point of severe wear.

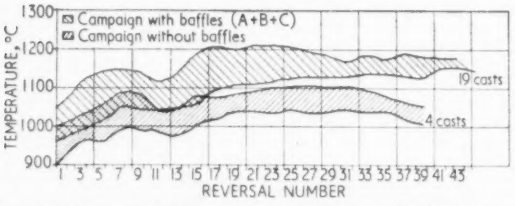
An expansion of 1 in. was left between the top of the deflector and the underside of the chamber arch. In service this gap was rarely taken up, so that eddies and regions of high local velocity caused by the gaps cut 'potholes' in the first ring of the chamber roof, which had to be removed at repair time. To avoid this the deflectors are now stopped 6 in. short of the chamber arch, and this appears to have remedied the 'potholing'.

PREHEAT IMPROVEMENT

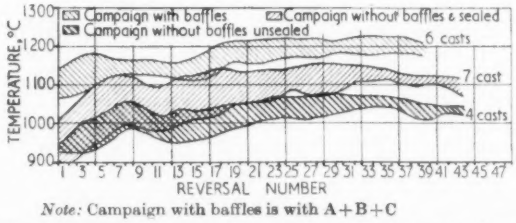
The scatter bands in Figs.18–21 show the preheat levels before, and after, inclusion of the baffles on four different single flue/checker furnaces.

Figure 18 is for *N* furnace, having flue splitters and bridgewall deflectors only. It can be seen that there is a clear improvement in all parts of the band. The maximum increase, between 110° and 130°C, occurs during the charging period. During melting and refining the increase lies between 70° and 90°C. Maximum average temperature is at 1180°C as against 1080°C during the previous campaigns.

Figure 19 for *M* furnace, with checker top baffles as well as the two features included on *N* furnace, show a similar range of preheat increase, but at a higher level. In this instance the charging period increase lies between 70° and 80°C, while that during refining is

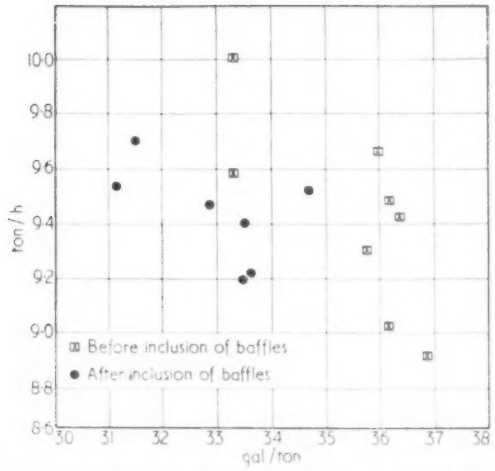


19 *'M' furnace. Mid reversal temperatures. Preheat scatter bands*

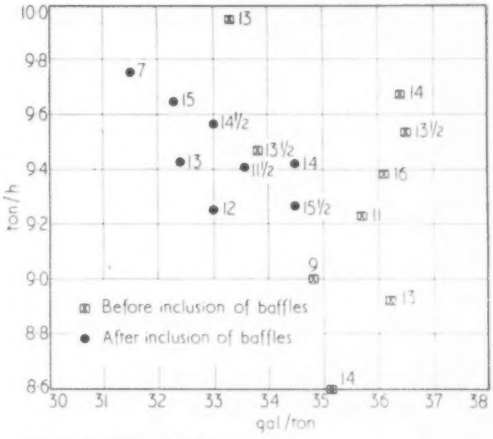


Note: Campaign with baffles is with A+B+C

21 *'S' furnace. Mid reversal temperatures. Preheat scatter bands*



22 *Furnace performance, single checker, single flue furnaces only eleven weeks,*



Figures denote weeks life

24 *Furnace performance, full campaigns. Single checker, single flue furnaces only, 14 weeks*

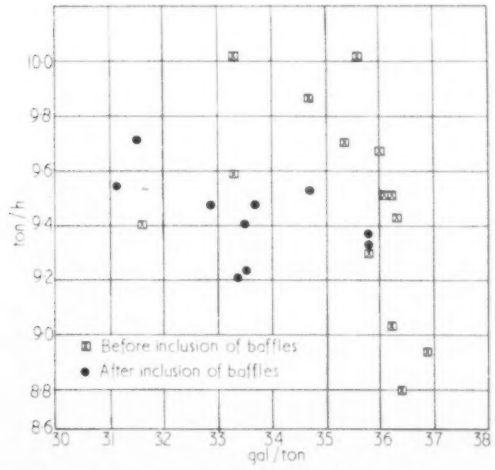
80°–110°C. The maximum average increase is from 1120°C before to 1220°C after the inclusion of baffles. A second campaign on this furnace, with the same modifications, provided even larger increases, i.e. a maximum average of 1245°C towards the end of refining. In addition, there was only a small period during the first and second charging runs, when the upper band line fell below 1200°C. It is thought that a difference in furnace operation during this second campaign (the steam/oil ratio was altered from the usual 8:1 to $5\frac{1}{2}$:1) was a factor in helping the baffles to last the whole campaign of 13 weeks.

Comparisons of *P* furnace campaigns in Fig.20 are slightly different. Here it is shown between preheats obtained at 425 gal/h with the original straighteners only, and those obtained with A+B+C. Improvement in maximum average is from 1160° to 1250°C. Increases of mid mean reversal temperature range from 30°C during charging, to 90°C during refining.

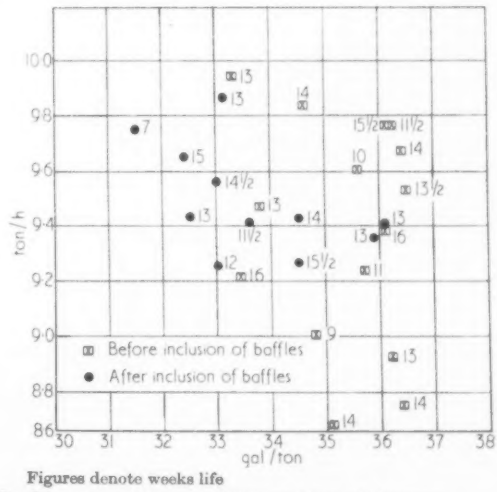
The three bands in Fig.21 were taken from three campaigns on *S* furnace; in effect they cover the whole survey, from the unsealed furnace to the sealed and baffled design. Of the bands, the lower is that obtained at the beginning of the survey, when general levels were measured, its highest point being 1070°C. In the centre is the band obtained when the furnace had been rebuilt, with the slag pocket air gap re-designed and all wickets sealed. At 1150°C the maximum average is then 80°C higher than the unsealed condition.

Finally, the top band, obtained when the baffles were used, shows a maximum average at 1230°C, with more than half the cast length above 1200°C. From the lower to the higher, the total improvement, taking highest levels only, is 160°C.

Regarding the upper bands in Figs.19–21, it can be seen that the trend on *M*, *P*, and *S*, is for the higher levels to be reached fairly early in the cast, and more or less held until tap. This is not the case for *N* furnace, where the preheat generally increases from finished charging until well into the refining period. It is thought that the checker top baffles are responsible for this, rather than effecting any further increase in the maximum preheat level.



23 *Furnace performance, all furnaces, eleven weeks*



Figures denote weeks life

25 *Furnace performance, full campaigns, all furnaces*

TABLE III Output and fuel rates

11 full weeks single checker/flue furnaces only		
	tons/h	gal/ton
Before (8 campaigns)	9·43	35·49
After (7 campaigns)	9·43	33·07
Difference	0·00	2·42
11 full weeks double checker/flue furnaces only		
	tons/h	gal/ton
Before (6 campaigns)	9·57	34·97
After (3 campaigns)	9·56	34·67
Difference	0·01	0·30
11 full weeks all furnaces		
	tons/h	gal/ton
Before (14 campaigns)	9·49	35·27
After (10 campaigns)	9·47	33·55
Difference	0·02	1·72
Full campaigns single checker/flue furnaces only		
	tons/h	gal/ton
Before (9 campaigns)	9·50	35·50
After (8 campaigns)	9·48	33·20
Difference	0·02	2·30
Full campaigns double checker/flue furnaces only		
	tons/h	gal/ton
Before (6 campaigns)	9·62	35·40
After (3 campaigns)	9·55	35·22
Difference	0·07	0·18
Full campaigns all furnaces		
	tons/h	gal/ton
Before (15 campaigns)	9·55	35·46
After (11 campaigns)	9·50	33·75
Difference	0·05	1·71

FURNACE PERFORMANCE DATA

The data used for Figs.22–25 consists of 26 campaigns from seven furnaces; four single checker/flue furnaces (17 campaigns) and three double checker/flue furnaces (9 campaigns). During the collection of the data all these furnaces were fired at a maximum oil rate of 425 gal/h.

The time period of collection ranges from March 1958 to December 1959, and covers conditions created by a low level of working (7–9 furnaces, mid 1958) and a high level (11–12 furnaces, mid 1959). It is considered that, during the low level of working, the performance of the individual furnaces improved, and that this is the only factor that has not been repeated at least once during the 26 campaigns. Therefore the 26 campaigns are considered sufficient to allow conclusions to be drawn on the effect of baffles and guides on furnace output and fuel consumption.

Only the first 11 full weeks have been taken for Figs.22 and 23, to provide a constant time basis for comparison. Because it could be argued that the baffles may be detrimental to furnace performance, if and when they collapse when nearing the end of the furnace campaign, Figs.24 and 25 have been included, showing the performance after full campaigns ranging from 7 to 16 weeks. The plots in the figures show a typical OH scatter, but two distinct groups occur. One group includes the furnace campaign after inclusion of baffles.

The plots shown in Figs.22 and 25 have also been divided into those concerning single checker/flue furnaces only, and those concerning all trial furnaces, both single and double checker systems. This is because insufficient data is available to warrant a separate statistical treatment for the double system results; a situation created by an alteration to all basic design of two of the three available double checker systems, after one campaign had been run in each furnace, with the baffles included.

From a statistical analysis of the data used in Figs.22 and 24 and given in Table III, the reduction in fuel consumption for the single system was found to be highly significant.

CONCLUSIONS

1. Prevention of air inleakage is an important factor in the attainment of the highest possible preheat for any given design and any mode of operation. A reduction of 80000 ft³/h inleaking air is accompanied by an increase in preheat of about 80°C.

2. Under normal design and operating conditions, the double checker/flue system tends to provide a higher preheat than the single checker/flue system, even though the filling volume of the latter is 16% greater than that of the former. The results of the trial show that there would be little or no improvement in air preheat on changing from a double to a single system, unless full use of baffles and guides were made.

3. Air preheat increases steadily from the beginning of a furnace campaign until it reaches a maximum at about half life. Thereafter it falls progressively as the furnace ages. This trend is attributed, in the main, to variations of air inleakage with furnace age.

4. Cold air tests in the single checker/flue furnaces showed that a poor distribution of incoming air, and outgoing waste gases, is a feature of normal design. Further tests in these furnaces showed that marked improvement of distribution was obtainable using baffles and guides. The normal distribution on double checker/flue furnaces was better than on the single checker/flue systems, and the addition of baffles and guides to the double systems made only minor improvement.

5. The improvement of preheat on single checker/flue furnaces was of the order of 110°C.

6. The increase of preheat obtained by the use of baffles and guides was manifest in a reduction of fuel/ton with a negligible effect on output rate. The reduction of fuel/ton on single checker/flue furnaces was 2·3 gal/ton or 6·7%. When both single and double systems are grouped together for analysis, the reduction in fuel consumption is 1·7 gal/ton or 4·8%.

7. On the single checker/flue furnaces the use of baffles appears to reduce the amount of roof repair owing, it is thought, to rectification of the incoming air bias in the uptake by the bridgewall deflectors.

ACKNOWLEDGMENTS

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Some observations on the interaction of liquid sodium with cast irons and plain carbon steels

A. A. Smith, M.A., Ph.D., and G. C. Smith, M.A.

SYNOPSIS

This investigation is concerned with the decarburizing action and allied effects of molten sodium on ferrous materials. It is shown that decarburization by sodium depends on the presence of a 'sink' in which the carbon removed from the specimen may be deposited. Cast irons are rapidly penetrated by the sodium and modification of the microstructure in the interior of the specimen may result, even where there is no removal of carbon from the iron. The formation of surface layers and crystals has been noted on steels immersed in sodium and these are ascribed to the transport of material from the container in which the test is carried out. Some measurements have also been made of the rate of decarburization of plain carbon steels by sodium over the temperature range 600–850° C, and the results correlated with the iron–carbon phase diagram and the diffusion rates of carbon in ferrite and austenite. 1732

INTRODUCTION

PUBLISHED WORK on the interaction of molten sodium with other metals has been largely confined to corrosion rates,[1, 2] with little discussion on the nature or possible mechanisms of attack. One of the first uses of molten sodium was as a coolant in aircraft engine valve stems, and although the sodium probably contained considerable amounts of oxide, so far as is known, no trouble was experienced from excessive corrosion. However, investigations have demonstrated that the attack of sodium on iron may be greatly enhanced by the presence of small quantities of oxide.[3] Besides direct attack, sodium may also decarburize plain carbon steels.[1, 4] The extent of this attack is reported to vary with the ratio of the volume of molten sodium present to the surface area of metal exposed, and sodium containing a high percentage of carbon may even carburize stainless steels. Precise evaluation of much of the information on the attack occurring on steels is further complicated by the fact that many of the results refer to attack under dynamic conditions in forced circulation loops. Apart from the fundamental interaction of sodium with the material of which the loop is constructed, attack under these conditions can also occur by cavitation and by mass transfer down a thermal gradient.

The present work has been restricted to a study of the attack of sodium on cast irons and plain carbon steels under static conditions, in an attempt to find out how sodium removes carbon from such materials and to investigate the factors affecting this form of interaction.

EXPERIMENTAL

Technique and Apparatus

The experimental method consisted of immersing and testing specimens in a known quantity of sodium contained in a sealed pot (Fig.1). The specimen pots were made of 18–8 stainless steel as this was reported to be relatively inert to sodium,[1] and the tests were normally carried out *in vacuo*. The test specimens, roughly $\frac{1}{4}$ in. thick, were cut from $\frac{3}{4}$-in. dia. rod (cast irons) or $\frac{1}{2}$-in. dia. rod (steels). They were polished metallographically on one face and lightly etched with 2% nital before placing in the specimen pots with the polished and etched face uppermost. The apparatus and technique for loading the specimens and sodium into the pots have been described in detail elsewhere,[5] and a schematic diagram of the apparatus is shown in Fig.2. The sodium was melted, degassed, and passed through a stainless-steel filter of 20 μ pore size before being admitted under vacuum to the specimen pot, which already contained the specimen. The lids of the specimen pots were ground to give a tight seating cone and socket joint with the pots (Fig.1) and could be pressed firmly home within the loading chamber. After removal of the specimen pot from the chamber, the lid was forced further in, and the joint between the lid and the pot autogenously welded to give a vacuum-tight steel. If the last two operations were performed with reasonable speed, it was found that no deterioration in the silvery white appearance of the sodium had occurred at the end of a test. This was taken to be the criterion of a satisfactory run, since it was not possible to measure the vacuum in the specimen pot once the lid had been placed in position in the loading chamber.

After the test the specimen pot was normally allowed to cool freely in air. The lid was then removed and the sodium was melted and poured out. The specimen was washed well in alcohol to dissolve traces of sodium and subsequently in distilled water and acetone before being dried in an oven. This treatment removed all traces of sodium from the steel specimens; in the case of cast-iron specimens, into which sodium

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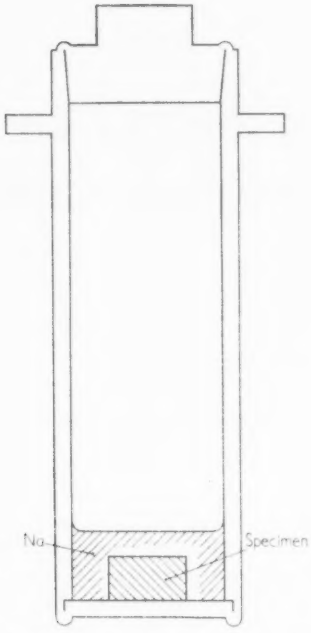
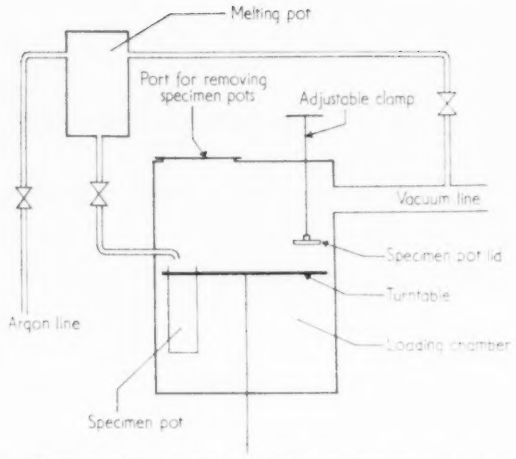


1 *Diagram of specimen pot, containing specimen and sodium*



2 *Diagram of apparatus for loading specimen pots with molten sodium* in vacuo

penetrated along the graphite flakes, the last traces of sodium were never removed.

The previously polished faces of the specimens were examined microscopically, and then sectioned. One half was kept for further surface examination, while the other half was chromium plated on the polished surface and examined in section microscopically. Table I gives an analysis of the materials used.

CAST IRONS

Roughening of the surface was so extensive during these tests that subsequent microscopical examination was impossible in spite of the fact that the specimens were polished before testing. They were therefore examined only in section.

One feature of the attack, peculiar to the graphitic cast irons, is penetration of sodium to the interior of the specimen along the graphite flakes. This was clearly demonstrated by 'sweating' of the graphite flakes in microsections of these specimens. Sweating is thought to be due to sodium, that has penetrated mechanically along the graphite flakes, reacting with the atmosphere to form sodium hydroxide and sodium carbonate. To overcome this, microsections were allowed to stand in water and then repolished. After repeating this treatment several times, 'sweating' was very much reduced. No 'sweating' was observed on the white cast irons or steels, and it is concluded that it is not in any way connected with diffusion of sodium through the metallic matrix.

Penetration of sodium along graphite flakes also resulted in erratic weight changes, making such measurements useless in estimating attack. Penetration is rapid; 5 h at 500°C and 2 h at 700°C enable sodium to penetrate to the centre of a specimen about 6 mm in thickness.

General features of the decarburization of pearlitic grey cast irons

500°C

Figure 3 shows a typical microstructure of a specimen treated at 500°C for 100 h; shorter times produced a similar appearance, but to a lesser depth. The light etching band at the edge of the specimen is seen under a higher magnification in Fig.4, and it can be seen that the structure in this region consists of unchanged graphite flakes embedded in a matrix which is partially ferritic and partially pearlitic, by contrast with the original matrix which was entirely pearlitic. Decarburization thus occurs by the partial removal of carbon from the matrix, the depth of the effect increasing with time of immersion. Partial decarburization tends to penetrate most deeply where there is a large number of graphite flakes close together, as can be seen in Fig.3. However, decarburization does not show any marked tendency to occur adjacent to the graphite flakes themselves (Fig.4).

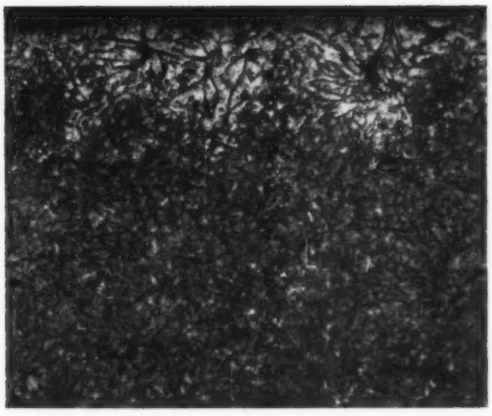
600°C

The effects at 600°C are very similar to those observed at 500°C but the decarburization rate is considerably faster. In addition there is a tendency for an edge

TABLE I Analysis of materials used

Material	C	Si	Mn	S	P	Cr	Ni	Cu
White cast iron	3·6	0·12	trace	0·015	0·02	—	—	—
Pearlitic grey cast iron	3·68	1·5	trace	0·016	0·02	—	—	—
Ferritic grey cast iron	3·55	2·90	trace	0·014	0·025	—	—	—
Hypoeutectoid steel *R*7	0·69	0·16	0·24	0·011	0·015	0·09	0·07	—
Hypoeutectoid steel *Z*	0·55	Not known						
Eutectoid steel *Y*	0·78	0·18	0·63	0·038	0·032	0·034	0·07	0·11
Hypereutectoid steel *K*	0·985	0·280	0·64	0·05	0·032	0·07	0·12	0·12
Hypereutectoid steel *W*	1·2	Not known						

There were also small amounts of W, Mo, Ti, Pb, and Al in steels *Y* and *K*.



3 *Section of grey cast iron showing decarburization after immersion at 500°C for 100 h* ×50

layer to form in which the matrix is completely ferritic, as seen in Fig.5.

An interesting pattern of decarburization was observed in a number of specimens heated for longer times, up to 100 h (Fig.6). Decarburization has occurred in such a way as to produce a pattern of ferrite regions which bear a striking resemblance to sections through dendrites. This 'dendritic' decarburization was most marked at 600°C. At lower temperatures the decarburization may not be sufficiently rapid for the effect to occur, while at higher temperatures, the very rapid decarburization results in a completely ferritic matrix except in those specimens heated for short times up to 2 h.

650°C

At 650°C similar effects were observed but they occurred more rapidly. In addition the pearlite remaining in the matrix showed a discontinuous appearance in that the carbide lamellae were broken up into a series of irregular short lengths. A similar effect was found in steels as shown in Fig.19.

4 *Detail of the edge region of the specimen shown in Fig.3* ×250

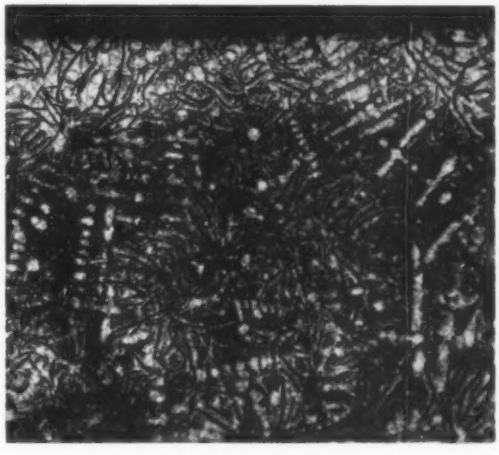
5 *Section of grey cast iron showing decarburization after immersion at 600°C for 5 h* ×250

700°C

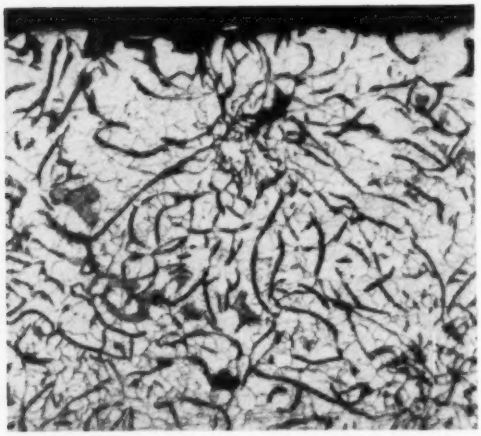
Specimens treated at 700°C showed no pearlite at all except in short experiments up to about 5 h (*see* Fig.7). A new form of attack became apparent, however, on specimens heated for longer times up to 100 h. This attack resulted in the entire or partial elimination of individual graphite flakes and spread slowly from the edge into the centre of the specimens with increasing times. Although it was not evident in all specimens, there was some tendency for the smaller graphite flakes to be attacked more rapidly than the larger ones.

750°C

At this temperature the attack on the graphite flakes became more severe and Fig.8 shows the structure in the transition region between attacked and unattacked flakes. The rate of decarburization of the matrix decreased slightly at 750°C as compared with 700°C; this is probably connected with the change of the matrix from α-iron to γ-iron. The appearance of the partially decarburized matrix was also somewhat different from that observed at lower temperatures;



6 *Section of grey cast iron showing 'dendritic' decarburization after immersion at 600°C for 100 h* ×500



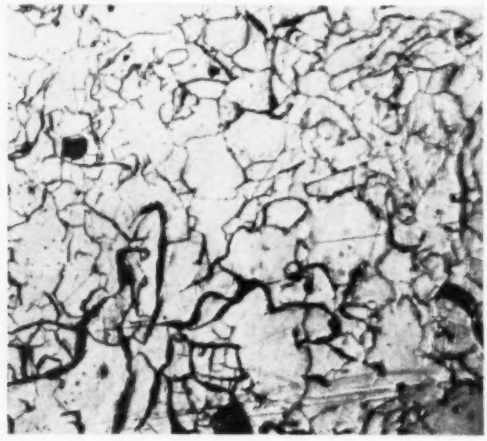
7 *Section of grey cast iron showing almost complete decarburization after immersion at 700°C for 2 h* ×100

there was a much greater tendency for decarburization to occur adjacent to the graphite flakes, as shown in Fig.9.

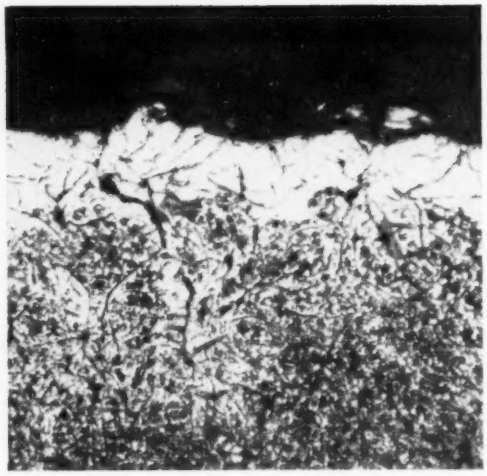
800°C

The features at 800°C were similar to those at 750°C, namely rapid attack on the graphite and a still further slowing down of the rate of decarburization of the matrix, such that even after 100 h treatment there was very little more decarburization than was seen in a specimen treated for 2 h at 750°C (*cf.* Figs.7, 9, and 10). In addition an edge layer of pearlite is seen in Fig.10 which was also seen to a lesser extent in specimens treated at 750°C, but not at lower temperatures.

With specimens treated at temperatures above 750°C the rate of cooling after the treatment had a marked effect on the final microstructure. Considerable pearlite remained if specimens were air cooled after treatment at 800°C; if, on the other hand, specimens were allowed to cool slowly in the furnace, complete decarburization of the matrix took place. This is attributed to the fact that during cooling the partially decarburized specimen spends a further 1½ h in the temperature range of fast decarburization, i.e. 750–600°C.



8 *Section of grey cast iron immersed at 750°C for 75 h; demonstrating solution of graphite flakes* ×250

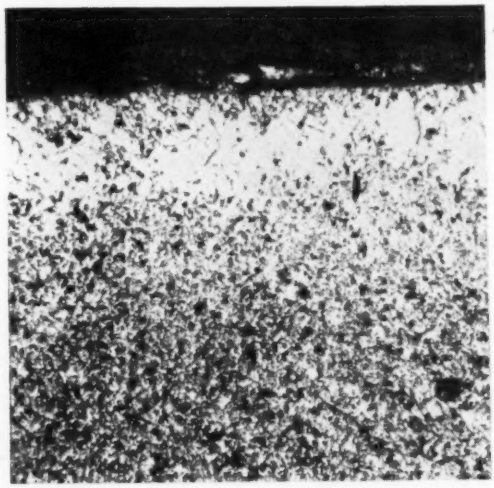


9 *Section of grey cast iron showing decarburization after immersion at 750°C for 2 h* ×100

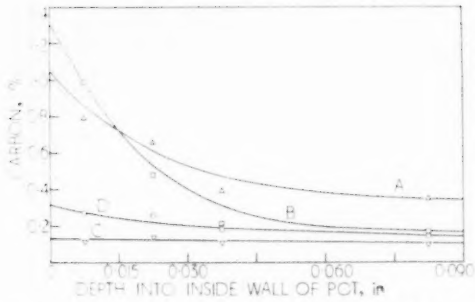
Additional experiments on pearlitic grey cast irons

Distribution of carbon after decarburization

Since the experimental system used was a sealed one, the carbon could not escape. This means that there are three possible explanations of the changes in carbon distribution during interaction. Carbon must dissolve in either the sodium or the container material or else remain in the specimen. The solubility of carbon in sodium containing 0·026% oxygen at 700°C is reported to be only 0·018%.[6] Facilities were not available for analysing sodium for carbon, but it was clear from the tests on steels in refractory lined pots (*see* 'Effect of the specimen pot', p.41) that the amount of carbon lost from the matrix could not all be dissolved in the sodium. As no undissolved carbon was found in the sodium it was possible that the carbon might have



10 *Section of grey cast iron showing decarburization after immersion at 800°C for 100 h* ×100



11 *Plot of carbon content* v. *distance from inside edge of specimen pot wall. Curves* A *and* B *below sodium level; curves* C *and* D *above sodium level*

been transferred to the specimen pot. Analyses were therefore obtained of the carbon content of specimen pots which had been used for several experiments on cast irons and steels. The results are shown in Fig.11, curves *A* and *C* referring to a pot which was used for three cast iron experiments and one experiment on a steel, while curves *B* and *D* refer to one cast iron and two steel experiments. The carbon content of the pots has been increased considerably in the regions in contact with liquid sodium; this carbon must come from the cast iron and steel specimens.

Effect of container material

Experiments were made in which the container was lined with refractory porcelain, so that the stainless steel was not in contact with the sodium. The results obtained with cast irons were similar to those obtained using unlined pots, in contrast with the results using steel specimens, where it was found that decarburization could be almost entirely suppressed by using a refractory lined pot.

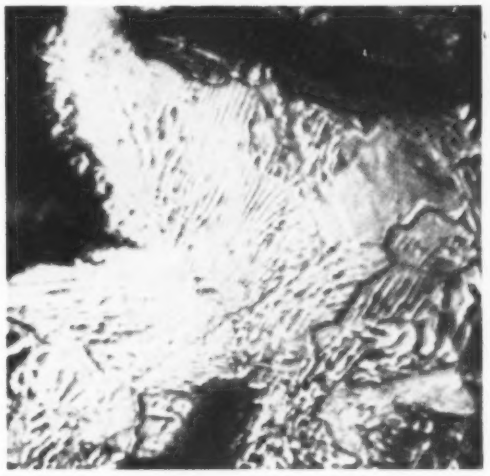
Appearance of pearlite regions after attack by sodium

Specimens which had been attacked by sodium at the lower temperatures, and in particular at 600°C, often contained ferrite regions which, under oblique illumination (Fig.12), showed a ridge structure reminiscent of pearlite but much less distinct than the appearance of true pearlite. Normal illumination showed a faintly striated appearance, but to see this clearly the specimen had to be etched in nital for a much longer time than was necessary to reveal the true pearlite.

To investigate this 'pseudo-pearlite' further, a specimen showing it to a marked extent was heated *in vacuo* in the austenitic range for about 15 min. A section now showed no 'pseudo-pearlite' at all, and the matrix was almost entirely ferritic, with just one or two regions of normal pearlite. This indicates that the pseudo-pearlite is not in fact ordinary pearlite but probably some ghost structure remaining after pearlite has been removed by the sodium decarburization. Additional confirmation of this was afforded by microhardness measurements taken on different regions of the microstructure. The following values were obtained:

Pearlite	330 DPN
Ferrite	190 DPN
Pseudo-pearlite	180 DPN

Examination in polarized light also showed that the pseudo-pearlite resembled ferrite more closely than



12 *Surface of grey cast iron, showing 'pseudo-pearlitic' markings after immersion at 500°C for 100 h* ×1500

pearlite. The extinction and maximum reflection positions typical of pearlite could not be obtained for pseudo-pearlite; instead the regions remained uniformly black, as with ferrite.

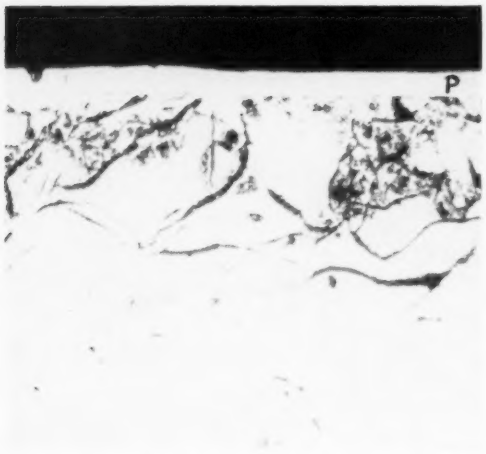
Factors affecting the appearance of edge pearlite formed at 800°C

Some specimens were quenched after treatment at 800°C to determine whether the changes in carbon distribution resulting in the formation of edge pearlite occurred during cooling or during the soaking period. Figure 13 shows that the increased carbon content at the edge was present even after rapid cooling. Structures having the appearance of graphite flakes near the free surface of the specimen may also be seen in Fig.13. In fact, these were voids which appeared to be the sites of former graphite flakes and which were opened up by the quenching; they were not observed on more slowly cooled specimens.

Further experiments showed that the soaking time at 800°C controlled the extent of the edge pearlite; it was not found in any specimens heated for less than 25 h, and usually 100 h were required to form a well-defined layer.

Oxygen contamination

In the work described so far the sodium used was as pure and free from contamination as possible. No quantitative measure of purity was available but the contamination was clearly not excessive, as judged by the untarnished appearance of the sodium after the experiments. To check the effect of oxygen some tests were carried out with sodium which had been deliberately contaminated. The results obtained were not markedly different with respect to form and extent of decarburization from those discussed earlier. The main differences were an enhanced attack on the graphite at any temperature, and a well defined band of pearlite separating the decarburized and undecarburized regions (Fig.14). In addition, at temperatures of 800°C, the grain boundaries of what were the austenite grains at the decarburizing temperature were clearly outlined by a series of dots which can be seen in Fig.15, penetrating slightly beyond the depth of the decarburized zone.



In this figure, also Figs.28 and 31, the plain white band at the edge of the specimen is a plating of chromium, deposited in order to preserve the edge regions during the sectioning and polishing of the specimens. This plating is marked *P* to distinguish it from the specimen proper

13 *Section of grey cast iron, showing edge band of martensite after immersion at 800°C for 100 h. Specimen quenched from testing temperature* ×250

Experiments with ferritic grey irons

In these irons no change in structure resulted from contact with sodium below 723°C. Above this temperature the effects were similar to those already described, in that edge pearlite was formed and direct attack took place on the graphite flakes. Another effect was the production of pearlite regions towards the centres of the specimens, but this was also shown by blank specimens heated and cooled *in vacuo*, and is not therefore due to the presence of the sodium. It is caused by graphite solution in the austenite matrix at 800°C and subsequent formation of pearlite on cooling.

Experiments on white cast irons

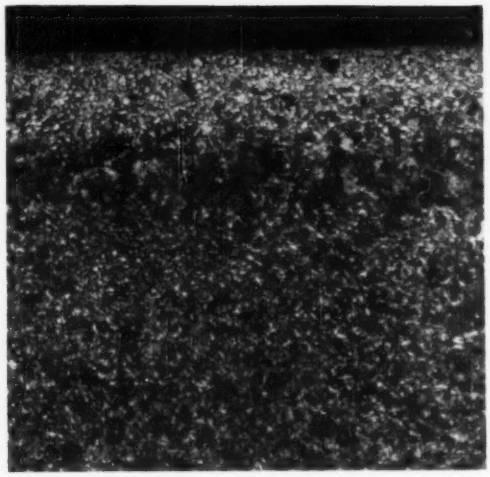
Interaction of white cast irons with sodium below the A_1 temperature produced slow decarburization resulting in an edge layer of ferrite, but no internal decarburization occurred. At 750°C a surface layer of ferrite was produced, and a well-defined layer of pearlite formed between the ferrite zone and the unchanged interior. A thin band of pearlite was also frequently observed at the extreme edge, similar to that observed on grey cast irons. Another interesting feature of the specimens was the frequent production of cracks in the decarburized region; the surfaces of these cracks also had pearlite layers analogous to the layer on the free surface (*see* Fig.16).

At 800°C there was no ferrite layer at the surface, but instead a quite broad band of pearlite containing cracks. At 850°C the cracks were less numerous, but the pearlite band was still observed, with a very thin layer of cementite on the free surface (Fig.17).

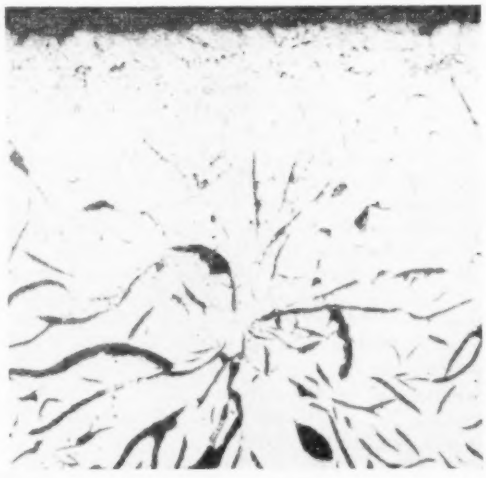
Discussion

Below the A_1 temperature, the structures developed by the grey cast irons can be explained in terms of direct decarburization and carbon redistribution. The sodium will remove carbon from the surface and transfer it to the specimen pot, and at the same time penetrate along the graphite flakes. The sodium present in the flakes is then effective in promoting a carbon redistribution in the matrix. It appears that carbon is removed from the matrix by dissociation of iron carbide, and by diffusion of carbon through the ferrite matrix to the sodium at the matrix/graphite interface. This carbon is then absorbed by the graphite so that the reaction continues, with the graphite flakes becoming larger as more carbon is removed from the matrix. Thus an apparent decarburization of the specimen occurs without carbon actually leaving the specimen. This internal decarburization is similar in principle to external decarburization promoted by the sodium, the only difference being that the carbon 'sink' in the former case is the graphite flake, and in the latter case the containing pot.

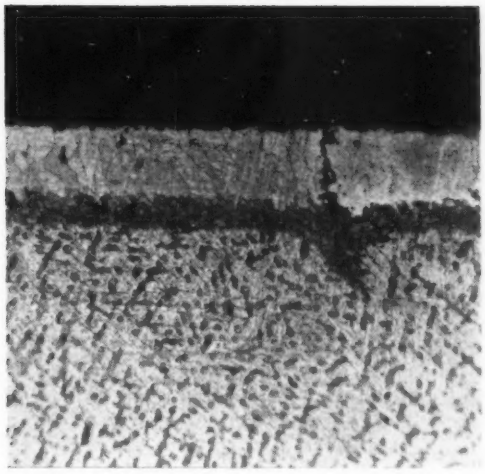
The partial decarburization structures developed at temperatures below 600°C are due to internal decarburization, the affected zone increasing in depth with



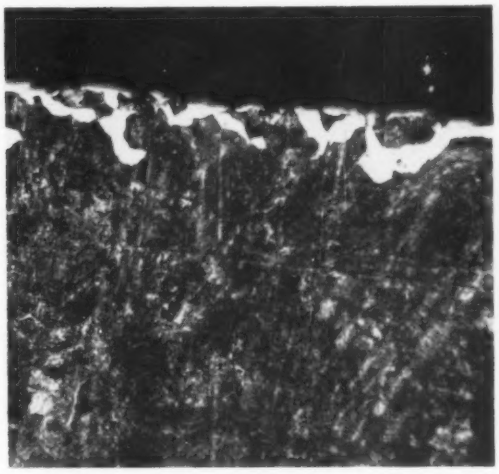
14 *Section of grey cast iron, showing decarburization after immersion in oxygen-contaminated sodium at 800°C for 25 h* ×75



15 *Same specimen as Fig.14, but unetched* ×150



16 *Section of white cast iron, showing decarburization after immersion at 750°C for 100 h. Note crack formation, which had pearlite following it to the edge of the specimen* ×50



17 *Section of white cast iron, showing formation of cementite at edge after immersion at 850°C for 100 h* ×200

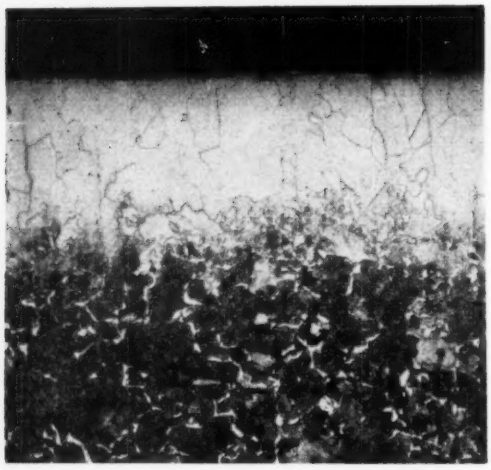
time. From observations on the 'sweating' of specimens, it appears that a certain minimum quantity of sodium is required in a graphite flake to promote internal decarburization, as 'sweating' could often be seen from graphite flakes nearer to the centre of a specimen than the partially decarburized zone.

The 'dendritic' type of decarburization is also due to sodium penetration along the graphite flakes. The exact mechanism is not clear, but it is probable that the effect is bound up with the microsegregation pattern developed during the solidification of the cast iron. The first-formed parts of a dendrite will tend to be richer in silicon than the parts that solidify later if equilibrium conditions are not maintained during cooling. It is known that the activity of carbon in steels is raised in the presence of silicon, and the activity gradient governing carbon diffusion during the internal decarburization will thus be greater from the silicon rich regions to the graphite than from the adjacent regions. This will thus promote a faster decarburization rate from the silicon rich areas, and hence produce a decarburized pattern identical with the original dendrite pattern. Eventually, however, as the time of decarburization is increased, all the interior carbon will be redistributed and the dendritic pattern will disappear.

The 'pseudo-pearlite' observed in decarburized regions may also be due to a microsegregation effect developed during the initial formation of the pearlite. The iron carbide in the pearlite will not contain the same concentration of impurity elements as the ferrite, so that when the carbide is removed by decarburization, the ferrite left is not of exactly the same composition as the original ferrite in the pearlite. The two regions may thus etch in slightly different ways although their hardnesses would be expected to be very similar.

Above the A_1 temperature, other factors may affect the microstructure. At such temperatures the cast iron consists of graphite flakes in a matrix of austenite saturated with carbon, to an amount which increases with temperatures up to 1130°C. Thus, as the temperature is raised above 723°C, some of the graphite dissolves in the austenite matrix, increasing the carbon content from 0·83%C to the saturation value at that temperature. The initial movement of the carbon is therefore the reverse of that occurring at lower temperatures. When sodium is present, it penetrates the remaining graphite flakes, as at lower temperatures, but does not now cause any internal decarburization of the matrix since the graphite and the carbon-saturated austenite are in equilibrium, and the carbon has the same activity in both phases. Decarburization can however still occur at the free surface of the specimen since carbon can be removed to the sodium and thence to the containing pot. This results in the matrix near the free surface being continually impoverished in carbon, and the graphite flakes in this region begin to dissolve to maintain the carbon content at the equilibrium value. In this way graphite flakes are eliminated to a depth which increases with time.

It is suggested that the formation of the edge layer of pearlite above the A_1 temperature may be due to a local change in the content of alloying elements at the free surface of the iron, caused by transfer through the sodium to or from the specimen pot. Changes of this nature, such as a decrease in the silicon content or an increase in the chromium content, will cause a decreased activity of carbon in the surface layers which may result in an increased carbon concentration, and yet still enable carbon to diffuse from the interior to the surface. Experiments by Darken[7] have shown that carbon will diffuse up a carbon concentration gradient from a high silicon to a low silicon steel. If the sodium in the present experiments dissolves some silicon from the iron, then boundary conditions similar to those used by Darken will be set up within the cast iron at the interface between the desiliconized rim and the interior of the specimen. If the rate of diffusion of carbon into the desiliconized zone is equal to or faster than the removal of carbon from the surface of the specimen in contact with the sodium, a surface layer relatively rich in carbon will be formed. A similar effect was noted by Harry,[8] who observed the appearance of pearlite colonies at the edge of a low carbon high silicon transformer iron, after the iron had been



18 *Plain carbon steel (0·69%C), sectioned to show decarburization after immersion at 675°C for 100 h* ×50

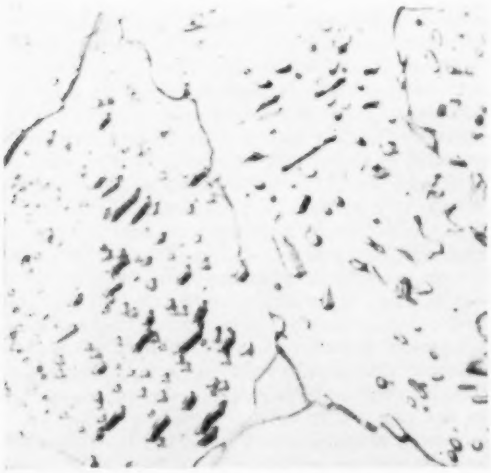
annealed in a faulty vacuum, which resulted in the removal of silicon near the surface by oxidation.

PLAIN CARBON STEELS

Introduction

The main interaction between steels and sodium is decarburization, as with cast irons. However, in steels this has been found to be purely an edge effect since there is no graphite network to allow internal decarburization. Effects noted on the surfaces of steels, which were not found with cast irons, included surface etching, formation of two types of crystals, raised grain boundaries, and the formation of a thin edge layer.

Steel specimens do not absorb sodium, and therefore their weight loss after interaction is a measure of the decarburization which has occurred. As with cast irons, the extent of decarburization is dependent on the nature of the containing pot.



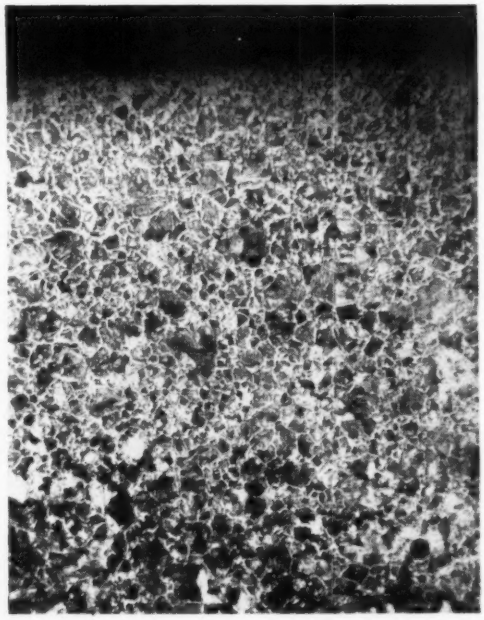
19 *0·69%C steel. Section showing detail of cementite in pearlite at the edge of the decarburized region after immersion at 720°C for 100 h* ×2000

20 *As for Fig.18, but immersed at 775°C for 100 h* ×50

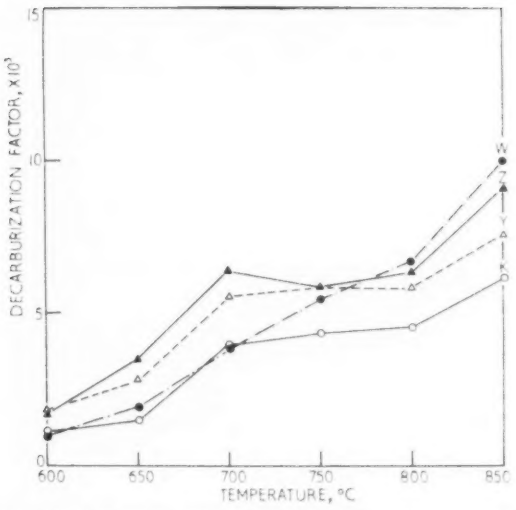
General features of the decarburization of steels

Temperatures up to 723°C

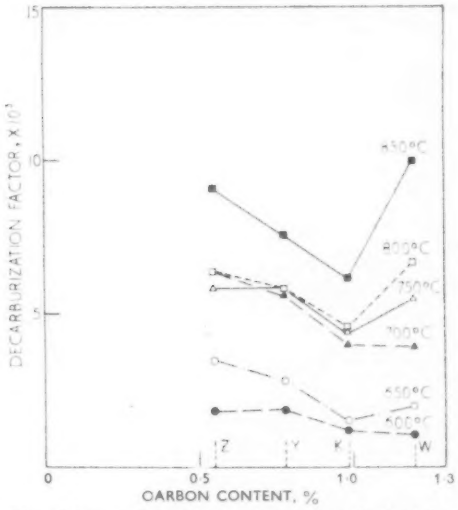
At 675°C a well-defined ferrite band was present with a fairly gradual transition to the original structure (Fig.18). The depth of this band of decarburization increased with temperature for any given steel, but decreased with increasing carbon content at any given temperature. The original pearlite beneath the decarburized band became very broken up and showed a



21 *As for Fig.18, but immersed at 850°C for 100 h* ×50



22 *Plot of degree of decarburization against the temperature of the test for steels Z, Y, K and W.*



23 *Plot of degree of decarburization against the initial carbon content for steels Z, Y, K and W at differing temperatures*

reduced amount of carbide (Fig.19). The grain on the left-hand side of this figure apparently shows sectioned needles rather than plates of carbide, resulting in an abnormal appearance of the pearlite. The ferrite grains in decarburized layers produced at temperatures from 675° to 723°C showed a gradually increasing tendency to columnar form, with increasing temperature.

Temperatures above 723°C

At these temperatures the matrix is more or less austenitic during decarburization, the relative amounts of austenite and ferrite, or carbide, depending upon the carbon content and temperature. The structure developed by contact with sodium was generally similar to that found at lower temperatures, but there were some differences. The demarcation line between the decarburized and unchanged regions was much more sharply defined than at lower temperatures, and the ferrite grains developed a pronounced columnar shape (Fig.20). No pearlite structure was found similar to that shown in Fig.19.

The structures developed by decarburization changed rapidly with temperature in the 0·69% carbon steel, owing to the small critical temperature range with this carbon content. At lower carbon contents, where the critical temperature range is much greater, the development of the marked columnar structure occurred more gradually as the temperature was raised above 723°C.

When the heating temperature was raised into the completely austenitic range a further change was observed. The well-defined ferrite layer was replaced by a continuous gradient of carbon content from the interior to the surface of the specimen (Fig.21).

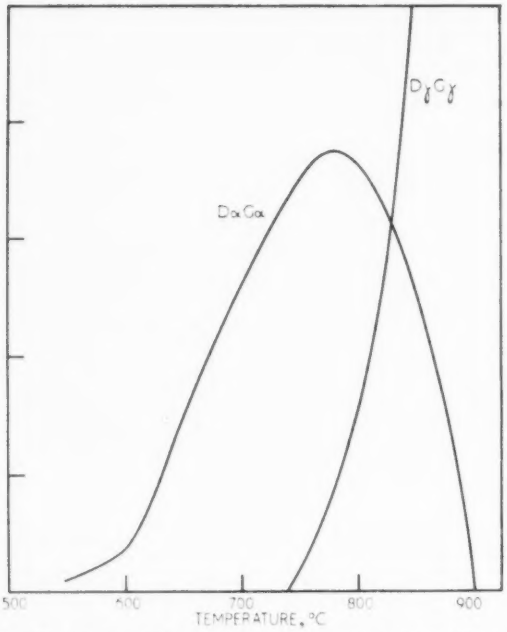
Weight changes as a measure of decarburization

The weight losses of steels of various carbon contents were determined after heating with sodium for 100 h in the temperature range 600–850°C. In contrast with the grey cast irons, the weight losses were reasonably consistent, owing to the non-absorption of sodium. It was assumed that weight changes were entirely due to decarburization, although this is not strictly true since small quantities of other elements may be lost or gained (*see* 'Effect of the specimen pot', p.41).

The decarburization curves are shown in Figs.22 and 23, the loss of carbon being plotted in terms of an arbitrary decarburization factor; this is defined as the weight loss of the specimen divided by the product of the initial carbon content and the surface area. All the specimens used were of similar size and the decarburization factor is in effect a measure of the depth to which appreciable changes in carbon content occurred. Values for the depth of decarburization obtained by calculation agreed reasonably with direct measurements from sections of specimens treated below the A_3 temperature; above this temperature direct measurements could not be made since there was no well-defined limit to the decarburized zone.

The extent of decarburization was found to rise with temperature up to about 700°C, and again beyond 800°C (Fig.22). The change with temperature in the range 700–800°C was much less marked and showed a reverse trend in the case of the lowest carbon content steels. The extent of decarburization also varied with carbon content, having a minimum value at about 1%C (Fig.23).

If it is assumed that the rate of dissociation of iron carbide and the rate of solution of carbon in sodium are both inherently high, then the rate of decarburization will be controlled by the rate at which carbon can diffuse in the steel. At temperatures below 723°C this diffusion will occur through a surface layer which is entirely ferritic for all carbon contents. At higher temperatures the structure of the surface will depend on the carbon content and the decarburizing temperature; thus a eutectoid steel at 800°C would be in the austenite range and the surface could lose carbon down to about 0·4% before becoming ferritic. On the other hand a 0·3%C steel would exist at 800°C as a mixture of ferrite and austenite and a lowering of the surface carbon content would result more easily in the formation of a ferrite layer. Whether or not such a layer forms depends upon the particular conditions of

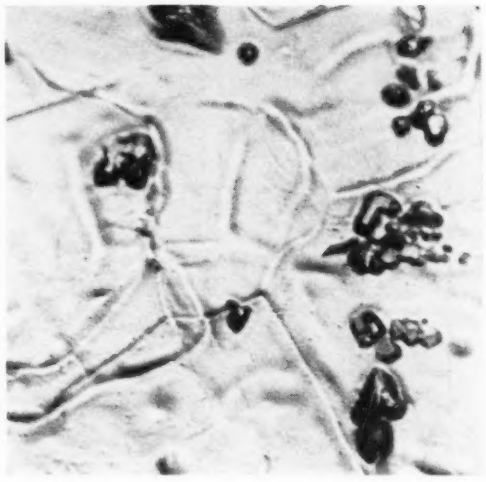


24 *The change with temperature of the product of the diffusion coefficient and maximum concentration gradient of carbon, for ferrite, and for austenite of eutectoid composition*

time and temperature for a given carbon content. In general it has been found that long times, low temperatures, and low-carbon contents favour the formation of such bands.

The rate of decarburization is difficult to predict theoretically owing to the number of factors involved; however, the following argument is suggested to explain the shape of the curves in Figs.22 and 23. For ferrite the diffusion coefficient of carbon increases exponentially with temperature, while the maximum carbon concentration gradient promoting diffusion (depending on the carbon solubility) increases to a maximum at 723°C and then decreases to zero at 910°C. The rate of transport of carbon at a given temperature will depend on the product of these two variables; Fig.24 shows how this product changes with temperature for ferrite and also for austenite. The curve for austenite shows no maximum since there is no maximum solubility reached for carbon in austenite in the temperature range of the present work.

As the temperature of decarburization is raised to 723°C, all carbon content steels should show an increased decarburization rate since carbon is being removed through a ferritic surface layer. Above 723°C the decarburization rate should slow down owing to the relatively low rate of carbon diffusion in austenite. However, at about 820°C the values of the products plotted in Fig.24 for ferrite and austenite become equal; above this temperature the slow step in a diffusion path will be diffusion through ferrite if such a band is formed at the surface. The microstructures show that this is not usually the case, but that austenite exists up to the surface for all carbon content steels for times up to 100 h. The decarburization rate should thus be controlled by carbon diffusion in austenite and increase rapidly with temperature.

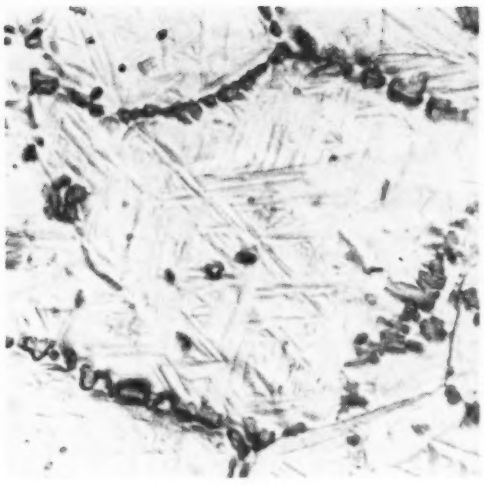


25 *Surface of 0·69%C steel, showing thermal etching and grey surface crystals after immersion at 800°C for 100 h* ×1000

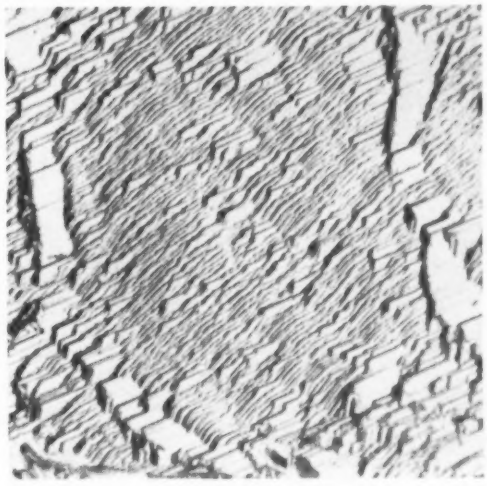
Two possible explanations may be put forward for the shapes of the curves shown in Fig.23. The first is that they are due entirely to differences in the contents of alloying elements in the steels. On this argument, steel *K* is always the slowest to decarburize because the minor alloying elements in it slow up the diffusion of carbon. Unfortunately detailed analyses of steels *Z* and *W* were not available to check this point. However, if composition is the important variable the fact that steel *K* has a markedly higher silicon content than steel *Y*, might suggest that steel *K* should decarburize the faster, which was not the case.

The second explanation is based on the structure of the steels. The general interpretation of Fig.23 is that the higher the initial carbon content up to 1%C, the slower the decarburization. Above 1%C the decarburization rate rises in experiments above the A_1 temperature, but appears to remain roughly constant for tests below the A_1 temperature. For hypoeutectoid steels tested below 723°C, the decarburization factor might be expected to decrease as the initial carbon content of the steel increases for the following reason. Initially decarburization of all pearlite grains in contact with the surface will take place at the same rate in all carbon content steels. However, after these surface grains have decarburized, there may be a considerable interval of ferrite in a low carbon steel which will enable pearlite grains some distance from the surface to start decarburizing. In a high-carbon steel the intervals of ferrite will be very small and so the depth of decarburization will increase more slowly. Thus, as the initial carbon content increases, the percentage of carbon removed in a given time decreases, although the weight of carbon removed increases. The curves for 600°C, 650°C, and 700°C thus show roughly the behaviour which might be expected, except that the minimum lies above the eutectoid carbon content. This might be due in part to the presence of minor alloying elements as previously suggested; it is difficult to see anything significant in the value of 1% carbon in itself.

Above the A_1 temperature the reason for the form



26 *Surface of 0·69%C steel, showing acicular etching and dark regions after immersion at 850°C for 100 h* ×1000



27 *Surface of 0·69%C steel, showing striational etching after immersion at 800°C for 100 h. Specimen quenched from testing temperature* ×1000

of the curves becomes obscure. At 750°C, there will still be sufficient ferrite present for the above argument to hold for the low carbon steels. The sharp rise in the decarburization factor for the steel *W*, and the persisting similarity of the shapes of the curves at higher temperatures are, however, puzzling.

The above discussion is concerned with the diffusion rate of carbon in the steel specimen. Under certain conditions, however, the diffusion rate of carbon in the specimen pot might be the rate determining factor for decarburization.

Surface effects visible after interaction of steels with sodium

Etching effects

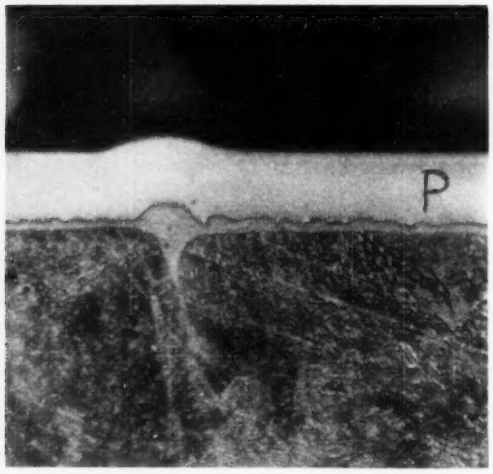
Three main forms of etching were found; these will be referred to as thermal etching, acicular etching, and striational etching.

The thermal etching shown in Fig.25 was observed only on specimens treated above the A_1 temperature and cooled slowly. The lines seen are due to the production of grooves at the austenite grain boundaries during the period of interaction. The fainter grooves are traces of former austenite grain boundaries which have disappeared as grain boundary migration occurred. This thermal etching is similar in appearance to the thermal etching which occurs on polished steel surfaces heated *in vacuo*, and is presumably due to the same cause, i.e. surface tension effects where grain boundaries meet a free surface.

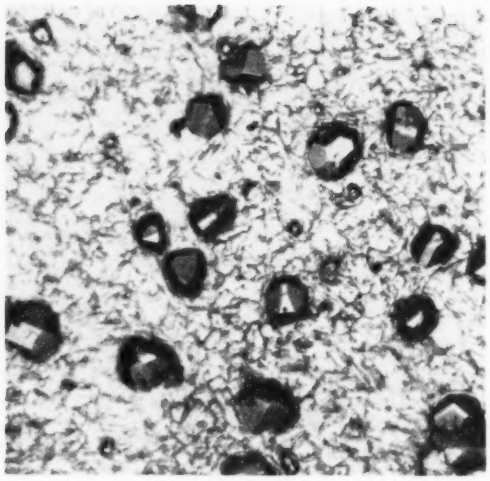
Acicular etching seldom occurred below 800°C. The structure, shown in Fig.26, is reminiscent of martensite in the shape of the needles and their relative orientations at about 45°, 60°, and 90°. On sectioning the specimens it was found that this type of surface coincided with the formation of a thin edge layer, which may be seen in Figs.28 and 31 (*see below* 'Additional surface effects'). This layer was so narrow that microhardness tests were of doubtful accuracy, although they indicated a high hardness. The appearance of the surface, its etching characteristics, its hardness, and comparison with other micrographs of polished and quenched steel surfaces[9] indicated beyond reasonable doubt that the layer was martensitic in nature.

The occurrence of a quenched surface layer on steel specimens which were air cooled must be due to some change in surface composition. It is suggested that transport of chromium from the specimen pot to the specimen (*see* 'Effect of the specimen pot', p.41) has taken place to a sufficient degree to render the surface air hardening. The fact that partially immersed specimens showed the layer only beneath the surface of the sodium also supports this hypothesis. In addition, the layer was suppressed by using a refractory lined pot, but this might be due to the slower cooling rate produced.

The striational form of etching, Fig.27, appeared on all specimens treated up to 800°C; at lower temperatures the etching was much finer, and at 650°C was only just visible on specimens treated for 100 h. Examination of sections suggested that, in contrast with the acicular etching, its formation depended on the absence of an edge layer. Thus, it is a possible form of etching when there is no mass transfer from pot to specimen. Because of the readiness with which the martensitic edge layer developed at 800°C and above, the striational etching showed itself in the main on ferritic surfaces; however, a few specimens showed very definite twin bands running across the striations, suggesting that the striations could form on austenite as well. The explanation of this type of etching presumably lies in the surface mobility of the iron atoms of the specimen when in contact with molten sodium, but the factors governing the size of the striations (which tended to increase in depth as the temperature rose) are not fully understood. It has been suggested that the thermal etching of silver in air is controlled by oxygen atoms,[10] and it may well be that, apart from the general effect of temperature, the presence of small amounts of oxygen in the sodium accounts for the considerable variation observed in the size of the striations. The banding effect visible in Fig.27 is thought to be due to slight grooves in the specimen surface.



28 *Section of hypereutectoid steel (about 1·3%C) showing fine edge layer and prominent ridge opposite grain boundary after immersion at 850°C for 100 h (see Fig.13)* ×1500



29 *Undersurface of 0·69%C steel, showing formation of large grey crystals after immersion at 850°C for 100 h* ×250

Additional surface effects

A further effect was visible on specimens air cooled after treatment at temperatures above 800°C (Fig.26). It consisted of small dark regions which appeared to follow grain boundaries and which were usually associated with the acicular etching. These are thought to be areas of nodular pearlite, formed when the cooling rate of the surface of the specimen or the build-up of chromium in the surface were not quite sufficient to result in a totally martensitic surface. The appearance of a partially immersed specimen agreed with this view. It exhibited a typical acicular etch below the sodium level and a smooth surface with slight grain boundary etching above the sodium. Just below the sodium surface there was a boundary region where the dark regions described above had formed copiously and extended downwards into the acicular region along the former austenite grain boundaries. This variation in microstructure can be correlated with the cooling conditions of the specimen. The part below the sodium level would cool faster than that above, and nodular pearlite may thus form at the austenitic grain boundaries just below the sodium level, where the cooling rate is intermediate.

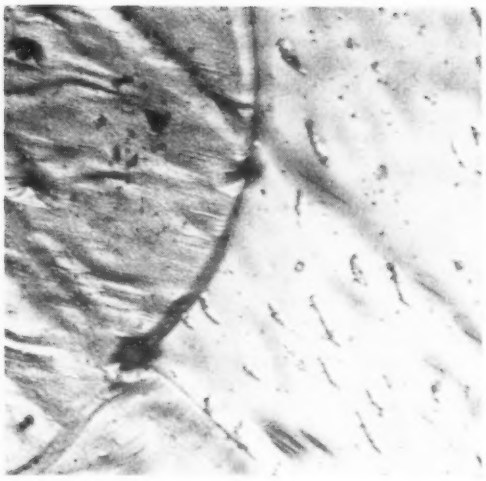
The surface layer associated with acicular etching sometimes developed a network of ridges 1–3 μ high, which resembled a grain boundary pattern in appearance. They formed above the termination of some grain boundaries at the surface (Figs.28 and 31). This effect is thought to be due to the transfer of material from the pot to the specimen. More alloying may occur at grain boundaries where diffusion rates are higher and this could result in an expansion of these regions with the formation of a ridge.

Surface crystals

Two types of crystal were found on the steel surfaces. One was large and grey, the other smaller and brilliant yellow. The large grey crystals, Fig.29, only occurred above treatment temperatures of 790°C, except when the specimen was furnace cooled, Fig.25. The crystals were found at random over the surface and showed no obvious orientation with respect to the underlying grains, or preferential nucleation at grain boundaries. They appeared only rarely, and no correlation could be found between their occurrence and the condition of the sodium, the type of specimen, or the pot. Normal etching reagents for steel had no effect upon them, and attempts at identification by electron diffraction were unsuccessful. The crystals bore no resemblance to the large crystal facets sometimes developed on metal surfaces immersed in liquid metals[4]; they appeared to have grown on the surface rather than to have been etched out of it. They were mainly found on the under surface of specimens, where the optimum conditions for pick-up of material from the pot would occur. It is thought that they are possibly an intermetallic compound resulting from mass transfer.

The small yellow crystals, Fig.30, usually appeared on the top surface rather than the under surface of a specimen; their occurrence varied consistently with temperature, since although they were few in number at 600–650°C, they were present on nearly all specimens, and above 750°C they were usually present in large numbers. In general, the higher the temperature the finer were the crystals. The yellow crystals in Fig.30 have grown considerably on the right-hand grain but on the left-hand grain they are still very small, suggesting that growth is determined by the orientation of the surface grains of the specimen. This was confirmed in specimens where crystals grew copiously on the untwinned portions of a former austenite grain but not on the twinned portions and vice versa. This also implies that growth of the crystals occurs on the specimen surface rather than that the ready formed crystals are precipitated from the sodium. They showed preferential nucleation in certain cases but this was not consistent. Thus, one specimen might show the crystals forming preferentially on ferrite as compared with pearlite, and another specimen might show the reverse. This indicated that the formation occurred during cooling, since the pearlite regions would not be present at treatment temperatures above 723°C.

Lining the specimen pot with refractory was found to suppress the crystals, but if a few flakes of



30 *Surface of 0·69%C steel, showing formation of small yellow crystals after immersion at 740°C for 100 h. Note marked preferential nucleation on the right-hand grain* ×1000

31 *Section of specimen shown in Figs.21 and 26 (immersed at 850°C for 100 h). Note thin edge layer and ridge formed at the left-hand side of the surface* (see *Fig.13*) ×1000

chromium metal were added to such a system, copious formation of the yellow crystals again occurred, indicating that chromium is a major constituent.

Specimens of copper heated under the same conditions as the steels did not show crystals, but specimens of Armco iron did, which indicated that iron was also essential for their formation, but probably not carbon. Other tests on deliberately contaminated sodium indicated that a certain amount of oxygen was also needed. The yellow crystals appear therefore to be a compound of chromium, iron, and oxygen. Variation in oxide content of the sodium may account for the varying numbers of the crystals seen on specimens. If contamination of sodium by the atmosphere was important in inducing the crystals to form, then it would be reasonable for the crystals to occur preferentially on the upper surface of the specimen.

Effect of the specimen pot

The specimen pots have been shown to have a marked effect upon the decarburization reaction by acting as a sink for the carbon removed from the specimens. In pots lined with refractory material, negligible decarburization of steels occurred under conditions which promoted oensiderable decarburization in unlined pots. The carbon removed from the specimens was not retained in the sodium, but taken up by the pot, as shown in Fig.11, the process being one of isothermal mass transfer of carbon. In view of this it was thought that transfer of other elements, such as chromium and nickel, might take place in the reverse direction, resulting in an enrichment of the surface layers of specimens.

Analysis of the surface layers of the pots were inconclusive in checking whether there was a decrease in the chromium, nickel, or silicon contents; neither was it possible to make a quantitative analysis of the surface layers of a specimen. However, qualitative analyses of the surface layers of several plain carbon steels treated with sodium in unlined pots showed a small but definite increase in chromium content; no increase was observed in nickel content. It appears, therefore, that isothermal mass transfer of chromium can take place under certain conditions.

Effect of liquid metals other than sodium

The liquid metals used were aluminium, bismuth, cadmium, calcium, lead, magnesium, tin, and zinc. Rigorous vacuum techniques were not possible when handling these metals, and therefore a higher degree of oxygen contamination was probable. With the exception of calcium no decarburization was observed, but extensive alloying had taken place during the tests with zinc, aluminium, and tin, and this might have obscured any signs of decarburization. The decarburization brought about by calcium on a 0·81%C steel after 100 h at 860°C was very patchy; this might be due either to incomplete melting of the calcium or to incomplete wetting of the steel surface by the calcium.

GENERAL CONCLUSIONS

The changes with temperature of the pattern of decarburization produced by sodium agree quite well with those obtained by the use of other decarburizing agents.[11,12] Sodium appears to act on steels in a similar way to other decarburizing agents, so that the decarburization structures produced by sodium may be predicted from the results of other decarburization studies. This does not apply to the decarburization of graphitic cast irons, where the interaction is much more complex owing to the penetration of sodium into the structure.

The second characteristic of sodium is that it acts as a bridge for the carbon during the decarburization process, rather than as a decarburizing agent in itself; in the absence of a 'sink' into which the carbon can be absorbed the sodium has slight decarburizing action.

The final degree of decarburization taking place in a sodium-filled system must depend on the level of carbon activity in the various components of the system, the ultimate distribution of the carbon being such that its activity levels are all equal. Since sodium may enable the transfer of other elements to take place,

this argument does not apply to the movement of carbon alone but to other alloying elements as well, the movement of which will also affect the carbon activity. Thus, rearrangement of elements will continue until each has a uniform activity level.

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Second-stage graphitization in a nodular cast iron

J. Burke

INTRODUCTION

THE EQUILIBRIUM FORM of most cast irons at room temperature consists of graphite and ferrite. Formation of this structure takes place in two stages: (*a*) graphitization of the carbon in excess of that soluble in austenite during solidification and/or annealing in the austenite range, referred to as 'first-stage graphitization', and (*b*) graphitization of the carbon soluble in austenite, termed 'second-stage graphitization'. In this second stage precipitation can occur directly from austenite during cooling or, alternatively, the austenite can first transform to pearlite, bainite, or martensite, which subsequently can be decomposed to graphite and ferrite by annealing below the eutectoid temperature.

The graphitization of pearlite and austenite normally proceeds by the further growth of the primary graphite,[1] whereas the decomposition of martensite results in the formation of a new generation of nodules.[2–4] This latter observation is surprising since it implies the existence of sites which nucleate graphite more effectively than graphite itself. Danko and Libsch[5] suggested that the new graphite forms by the decomposition of an unstable silicon-rich carbide produced during tempering of the martensite. Studies of

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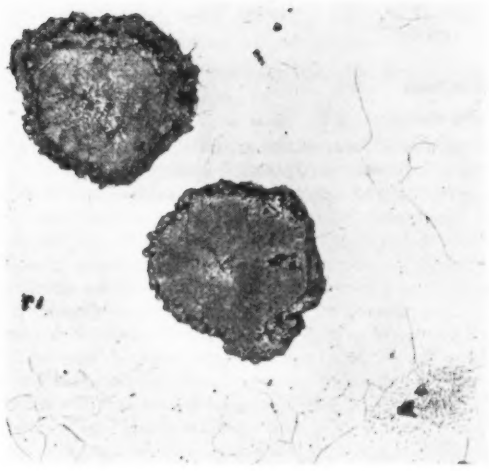
SYNOPSIS

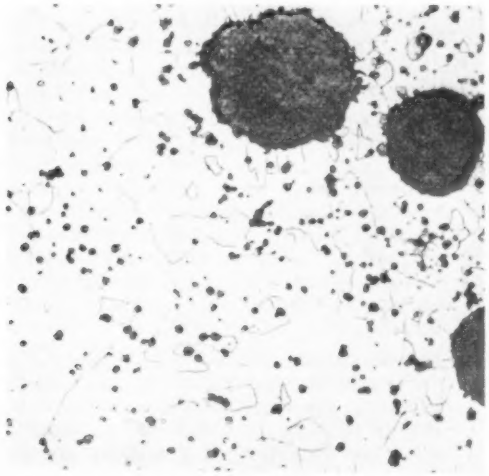
A study has been made of the kinetic and microstructural characteristics of the graphitization of pearlite and martensite in a nodular cast iron. The decomposition of martensite invariably resulted in the formation of new nodules of graphite; a similar structure was also observed occasionally in samples initially pearlitic. However, the graphitization of pearlite more often produced secondary graphite in the form of a deposit upon the primary nodules. The kinetic features indicate that before the primary nodules are capable of further growth at temperatures below the eutectoid a time-dependent nucleation process is necessary and that the rate of growth of graphite is diffusion controlled. 1836



1 *Pearlitic specimen graphitized at 690°C. Peripheral graphite is clearly distinguishable from the primary nodules. Etched 1% nital* ×300

silicon steels indicate that the carbide formed during the third stage of tempering contains silicon initially. Rejection of the silicon occurs at such a speed that at 600–700°C a few minutes is sufficient to produce pure cementite.[6] It appears unlikely that this transition carbide would have a direct influence on graphitization since the incubation period of the latter is 30–60 min. Furthermore, Morrogh[7] considered that the evidence presented by Danko and Libsch in support of this theory was not conclusive.

It is clear that more work is required before a satisfactory explanation of this important problem can be found. It was considered that a comparison of the kinetic and structural features of the graphitization of pearlite and martensite would provide useful data. A nodular cast iron was chosen for the investigation because the shape of the primary graphite facilitates the measurement of growth rates.



2 *A new generation of graphite nodules produced during secondary graphitization of pearlite* ×200

TABLE I Values of the index n

Temp., °C	Pearlite matrix Initial 40% graphitization	Pearlite matrix Final 60% graphitization	Martensite matrix Initial 40% graphitization	Martensite matrix Final 60% graphitization
710	2·44	1·45	3·24	1·87
690	2·30	1·50	3·44	1·61
670	2·20	1·52	3·54	1·76
Average	2·32	1·49	3·41	1·75

EXPERIMENTAL DETAILS

The nodular cast iron used was of commercial origin, made by the standard process of inoculation with a magnesium–nickel alloy and ferrosilicon, the composition being 3·3%C, 2·5%Si, and 2·0%Ni.* The casting was given a preliminary anneal at 900°C for 4 h to ensure the absence of primary carbides. One half of the specimens machined from this casting were reheated to 900°C for 30 min and cooled in a stream of cold air which produced a matrix of martensite; the remainder were given the same heat-treatment and then transformed isothermally at 600°C for 5 min, to produce fine pearlite devoid of free ferrite. All specimens were nickel-plated to prevent oxidation and decarburization during these and subsequent heat-treatments.

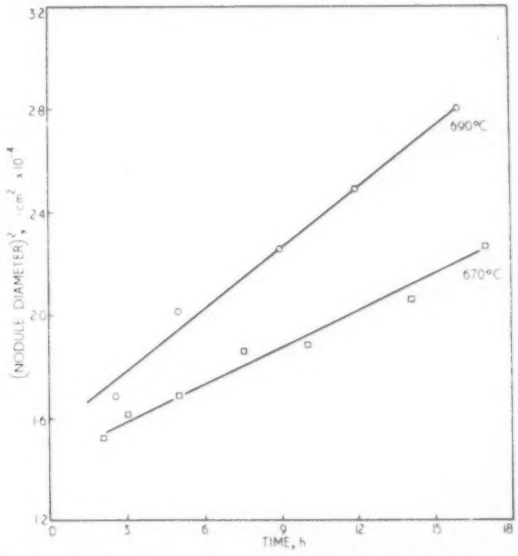
Isothermal secondary graphitization was studied as a function of initial structure at 670°, 690°, and 710°C using standard dilatometric and metallographic techniques.

RESULTS

Metallographic observations

The secondary graphite resulting from the decomposition of martensite invariably occurred as completely new nodules, distinct from primary graphite. Similar

* Other elements were not determined in this alloy. However, this source of material normally gives the following: 0·30%Mn, 0·04%S, 0·06%P, and 0·04%Mg.



3 *Graphs of the square of the diameter of the largest primary nodule as a function of time of annealing at the temperatures shown, for pearlitic specimens in which secondary graphite occurred as a deposit on primary nodules*

results have been reported by Rehder and others.[2–4] There was no noticeable variation of morphology of the secondary graphite with reaction temperature. The martensite tempered to a fine sorbitic structure before graphitization started. Heat-tinting of specimens annealed for 75% of the incubation period stained the carbide phase uniformly dark, suggesting that only cementite was present.[8]

In about 90% of the pearlitic samples the secondary graphite was found as a deposit on the primary nodules. The two types of graphite could be readily distinguished, suggesting that they differed in orientation or grain size (Fig.1). It was noted during metallographic preparation that, whereas it was relatively easy to retain the primary nodules intact, it was not possible to retain the peripheral secondary graphite in more than a small fraction of the nodules. These observations show that there was no coherency between the two graphites.

Deposition did not start on all nodules simultaneously, occurring first on intermediate sizes and later on the larger and smaller ones. All nodules were participating in the reaction by about 40% graphitization. This observation is important in relation to the kinetic results.

In the other 10% of the pearlitic samples the secondary graphite occurred as new nodules, the structure being similar to that of the martensitic specimens (Fig.2). This result was not confined to any particular temperature. It is believed that this is the first report of this structure in irons that were initially pearlitic. It shows that tempered martensite is not a necessary prerequisite for its formation.

Kinetic results

Both structures produced reaction-rate curves of the sigmoid type after a definite incubation period, the magnitude of which decreased as the temperature increased. Martensite graphitized more quickly than pearlite with a shorter incubation period; this is in accord with the results of Gittus.[9] It was not possible to measure the rate of decomposition of pearlite reacting to new nodules. However, it was apparent from metallographic observations that it was much faster than when existing nodules extended.

The isothermal rate curves (obtained dilatometrically) were analysed in terms of the equation

$$y = 1 - e^{-(t/k)^n} \quad \ldots\ldots(1)$$

in which y is the fraction of the secondary graphite formed at time t, and k and n are constants.[10] The values of n changed at about 40% reaction. The actual values are given in Table I.

It was possible to measure the rate of radial growth of graphite only in those pearlite specimens in which deposition on primary nodules occurred. The largest diameter of the largest primary nodule was measured microscopically in a series of partially reacted samples and plotted as a function of annealing time. On the assumption that the patch measured was an equatorial section of the largest primary nodule and that the size of this nodule was the same in each sample before secondary graphitization, then the gradient of a graph of diameter against time is twice the radial rate of growth of the nodule. An approximately parabolic relationship was found between the nodule diameter and the time. This is shown in Fig.3 for the results at 670° and 690°C.

DISCUSSION

Kinetic results

The parabolic form of the growth curves indicate that the rate of growth is controlled by a diffusion process.[11] However, it was not possible to establish which diffusion process is the important one. For reactions proceeding by the diffusion-limited growth of spherical particles the theoretical value of the index n in equation (1) is 1·5.[12,13] If nucleation also occurs, then this value is increased by the value of the time exponent in the expression giving the number of nodules per unit volume as a function of time, e.g. if the nodule number is directly proportional to the time n is increased by unity.[14] In the pearlitic samples all the primary nodules did not start to grow together; this indicates that a nucleation process occurs progressively during the first 40% reaction. The increased value of n in the first part of the reaction is attributed to this process. After nucleation ceases the transformation involves growth only; the final value of 1·49 is in good agreement with the theoretical value. There is also a nucleation process involved in the martensitic irons. By analogy with the graphitization of cementite in the austenite range[10] it may be presumed that this ceases at about the half-way point. The different values of the index can be accounted for on this basis. The final value of 1·75 which refers to a period when growth only is operative, is higher than the theoretical value of 1·5. This discrepancy is attributed to departure from spherical geometry which increases n.[12]

Nucleation of secondary graphite

The presence of an incubation period, normally associated with the time required to form a stable nucleus, and the fact that new nodules are often formed as a result of secondary graphitization, demonstrates that the primary nodules are incapable of further growth without a nucleation event. Two possible reasons for this are:

(i) the primary nodules are coated with impurity during the heat-treatment preceding secondary graphitization. A few atomic layers of impurity would be sufficient to effectively isolate the graphite from the matrix, in which case the primary nodules function merely as large inclusions upon which heterogeneous nucleation is possible. The orientation of secondary graphite deposited upon these 'inclusions' should be random unless there exists an orientation which leads to a low-energy interface between the graphite and impurity; in the latter case it is to be expected that the orientation and grain size of primary and secondary graphite would be identical

(ii) the dislocations responsible for primary growth are rendered inoperative, by, for example, thermally generated stresses or impurity locking. In this case growth would continue only after the formation of a suitably dislocated nucleus on the surface. A similar orientation of primary and secondary graphite is to be expected in this case since any other would involve

the formation of a graphite/graphite interface of non-zero energy. Thus the two types of graphite would be indistinguishable.

The present results indicate that the first explanation is the more probable.

The formation of new nodules during secondary graphitization can now be more readily understood since it simply requires the existence of sites at which the activation energy for nucleation of graphite is less than at these 'inclusions'. The occurrence of new nodules in pearlitic irons is taken to indicate that a silico-carbide is not necessary for the formation of this structure. During the incubation period preceding graphitization pearlite spheroidizes and martensite tempers so that when graphitization starts both types of specimen consist of particles of cementite in a ferrite matrix. It is considered that new nodules are formed at the interface between the cementite and ferrite.

The change in free energy, ΔG accompanying the formation of a nucleus of graphite of volume, v and surface area a_1 is given by

$$\Delta G = v(\Delta G_c + \chi) + a_1\gamma_1 - a_2\gamma_2 \quad \ldots\ldots(2)$$

where ΔG_c is the change in chemical free energy per unit volume of graphite, χ is a parameter measuring the strain energy involved, γ_1 is the specific surface free energy of the graphite/ferrite interface, γ_2 is the specific surface energy of the surface at which nucleation occurs, and a_2 is the area of this surface destroyed by the formation of the nucleus. ΔG_c is negative and all other terms positive. Nucleation occurs at those sites at which ΔG is least. It is reasonable to assume that the strain energy will be the same irrespective of where the nucleus forms. Since it is differences in ΔG which are of interest, the strain term may be ignored. Similarly a_1 γ_1 need not be considered. There is no reason to think that ΔG_c is a function of position in the pearlitic samples; thus γ_2 is the determining factor. The fact that secondary graphitization usually occurs at the primary nodules in pearlitic iron indicates that the cementite/ferrite interfacial energy is less than that associated with the interface between ferrite and the impurity surrounding the graphite nodules. Special conditions, occasionally realized in this work (*see* below) are necessary to reverse this behaviour.

Conditions in the martensitic samples are different. The carbide formed initially in the third stage of tempering contains silicon[6, 15] which is later rejected. However, since the diffusion of silicon in ferrite is slow at these temperatures there will be an excess of this element in regions adjacent to cementite particles. Silicon increases the activity of carbon in ferrite; thus carbon supersaturation relative to graphite is increased in these regions, i.e. ΔG_c is increased. A second important fact is that the size of the cementite particles is much smaller in tempered martensite than in spheroidized pearlite. This has two consequences, (1) ΔG_c for nucleation at the ferrite/cementite interface is increased because the concentration of carbon in ferrite coexisting with cementite increases as the particle size of the latter decreases, and (2) the specific surface energy of the carbide particle is increased (i.e. γ_2). All three factors favour nucleation at the cementite/ferrite interface in preference to the primary nodules and their combined influence should be sufficient to ensure that the former are the favoured nucleation sites in tempered martensite.

It is of interest to estimate the difference in the activation energy for nucleation in martensitic and pearlitic samples (and thus at the cementite/ferrite and nodule/ferrite interface). This is obtained by comparison of the incubation periods, t, given by

$$t = Ae^{+Q/RT} \quad \ldots\ldots(3)$$

where Q is the activation energy for nucleation, A is a constant, and the other symbols have the usual meaning. At 710°C the incubation period for the martensitic specimens was 1/5th of that in the pearlitic irons, equivalent to a difference in Q of 3200 cal/mole. Taking the activation energy for nucleation for first-stage graphitization as an indication of the probable order of magnitude of Q,[10] this value represents only a 5–6% difference. It is reasonable to suppose that the three factors discussed above can effect such a small change in Q and thereby change the structure. Also it is unlikely that the difference could be detected by present nucleation counting methods.

It remains to consider the formation of new nodules in those pearlitic samples found to yield this structure. As discussed above, γ_2 is the important factor in these irons. New nodules imply that the cementite–ferrite interfacial energy is greater than the primary nodule–ferrite interfacial energy. Impurities are known to affect interfacial energies. It is possible that an unusual distribution of impurities occurred in those samples, causing a reversal in the relative magnitudes of the two interfacial energies.

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Investigations on the tuyere and combustion zones in blast-furnaces and low-shaft furnaces

Rolf Ebert and Karl-Friedrich Lüdemann

SYNOPSIS

Measurements of the depth of penetration of the tuyere zone, carried out on German blast-furnaces and low-shaft furnaces, confirm the rules found in work carried out in the UK in 1957 on an experimental furnace and on blast-furnaces. The nature and particle size of the coke are particularly important factors. The configuration of the combustion zone in front of the tuyeres in the low-shaft furnace is the same as the long-known configuration in the blast-furnace. There is no unique correlation between the extent of the combustion zone and the depth of the tuyere zone; the influence of blast temperature and exit velocity is nevertheless discernible. 1830

INTRODUCTION

AN IMPORTANT FEATURE of the process taking place in the blast-furnace and the low-shaft furnace, both of which draw their heat requirements from the combustion of carbon, is the use of large amounts of fuel mainly in the form of metallurgical coke. The majority of the coke burns in front of the tuyeres, within a limited space, in the medium of the hot blast injected from outside. Bearing in mind that the fuel may constitute 60% or more of the total volume of burden, the importance of the combustion process in controlling operation of the furnace is readily appreciated. Since 1893, when van Vloten[1] published the first results of his experimental investigation of the processes taking place in front of the blast-furnace tuyeres, considerable research has been carried out on this problem. The experimental methods used indicate two main lines of attack:

(i) investigation of the movement of the coke and the size of the tuyere zone in front of the tuyeres

(ii) investigation of the actual combustion process in front of the tuyeres by taking gas samples.

Observations and measurements have been made both on small experimental furnaces and on industrial blast-furnaces. Among the more recent research, particular importance has been attached to the measurements of Taylor *et al.*,[2] published in 1957, of the extent in space of the tuyere zone under various conditions. The measurements made on an experimental furnace indicated the relationship between the depth of penetration of the tuyere zone (axially to the tuyere) on the one hand, and the quantity of hot blast in unit time per tuyere, relative density of the hot blast, tuyere diameter, and particle size of the coke, on the other hand.

The formula proposed by Taylor *et al.* is as follows:

$$P = B \frac{V^{0.7}}{\rho^{0.2} D^{0.4} d^{0.5}} \quad \ldots\ldots(1)$$

where P is the depth of penetration of the tuyere zone in cm, B is a constant, V is the quantity of blast per tuyere, in Nm^3/min, ρ is the relative density of the hot blast, D is the tuyere diameter in mm, and d is the mean particle size of the coke in mm.

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The relative density of the blast used in this formula is the ratio of the density of the hot blast to that of air at +20°C and 1 atm pressure. Strictly speaking, d should refer to the particle size of the coke at the tuyere zone. Since in general this is unknown, use must be made of the mean particle size as determined before the coke is charged into the furnace.

The values determined by Taylor *et al.* on the experimental furnace agree well with equation (1). However, over a wide range of coke sizes there were certain discrepancies, obviously attributable to the fact that use had been made of the mean diameter of the coke particles, not of their mean weight. The depth of penetration of the tuyere zone in industrial blast-furnaces was measured using wooden rods, saturated with water, which were inserted through the tuyeres. Here again the predictions of equation (1) were by and large confirmed.

Works experiments

During 1958–59 measurements have been made by the Eisenhütten-Institut der Bergakademie Freiberg (Saxony) of the extent of the tuyere zone and its dependence on driving conditions in a series of German blast-furnaces and low-shaft furnaces, involving 385 series of readings and about 3600 individual measurements. It was found possible to obtain useful data with the simplest of means. The major characteristics of the furnaces examined, such as useful volume, type of hearth, diameter, and number of tuyeres and types of pig iron made, are collated in Table I.

The use of water-saturated wooden rods, of 20 mm square section, was again found to be most convenient.

TABLE I Characteristics of blast-furnaces and low-shaft furnaces examined

A Blast-furnaces

Works	Furnace no.	Useful vol., m³	Hearth dia., m	Tuyeres no.	nozzle dia., mm	Depth of penetration of tuyere zone, mm	Type of iron made
	I	670	6·10	12	200		Steelmaking
	II	662	5·60	10	220	75 to	Hematite
	III	645	5·60	10	220	110	Steelmaking
	IV	680	6·10	12	200		Steelmaking
S	V	680	6·10	12	200		Steelmaking
	VI	645	5·60	10	220		Steelmaking
	I	255	4·20	7	180		Bessemer
M	II	444	5·00	8	180	65 to	Bessemer
	III	436	5·00	9	180	110	Bessemer
	IV	446	5·00	9	180		Bessemer

B Low-shaft furnaces

Works	Furnace no.	Useful vol., m³	Hearth length, m	breadth, m	Tuyeres no.	nozzle dia., mm	Depth of penetration of tuyere zone, cm	Type of iron made
M	I	50	6·35	1·65	14	100	30–60	Spiegeleisen
C	IX	78	6·00	1·80	14	135	25–60	Foundry iron
	X	81·5	6·00	1·80	16	135	25–60	Foundry iron

Table II gives the individual readings obtained in measurements of the mean depth of penetration of the tuyere zone on blast-furnace I at works *M*. In spite of the relatively high scatter of the readings, between 69 and 106 cm, extraction of the arithmetic mean proved to be valid. The quantity of blast per tuyere, which must be regarded as the most important factor, could not be measured directly; in all cases it was derived from the known total blast input to the furnace and the number of tuyeres. The losses in pressure and quantity occurring between the measuring position and the tuyeres were not allowed for; exact determinations would have entailed considerable instrumentation. The magnitude of these losses was roughly the same for all the furnaces examined.

On furnace I at works *M*, measurements were made of the pressure difference at each tuyere, over the tuyere connexion, compared with a measuring position at the junction of the hot-blast main and the ring main. The technique has recently been described by Wolf *et al.*[3] The relationship between the pressure difference and the mean depth of penetration of the tuyere zone is shown in Fig.1 for tuyeres 1–7, with the furnace on low and normal blast input; each input is the mean of about 25 individual readings. Figure 1 indicates that the depth of the tuyere zone in front of each tuyere is primarily determined by the corresponding blast throughput, as directly indicated by the corresponding value of pressure difference. The relatively wide scatter of the points must be ascribed to causes associated with operating conditions; thus for example, tuyere no.1 is opposite the junction of the hot-blast main and the ring main, so that it exhibits the highest pressure drop *a priori*.

When the mean depth of penetration of the tuyere zone is plotted against the value

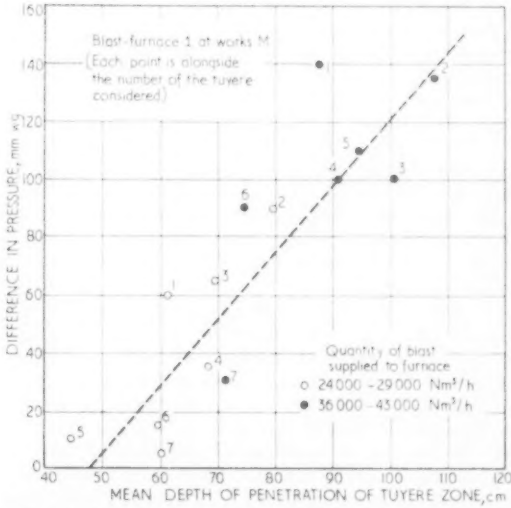
$$P = B\frac{V^{0.7}}{\rho^{0.2}D^{0.4}} \qquad \ldots\ldots(2)$$

derived from equation (2), for coke sizes of 20, 30, 40, 50, and 60 mm, the graph shown in Fig.2 is obtained. The values referring to UK blast-furnaces investigated by Taylor *et al.*[2] clearly show the influence of particle size. Thus, at a given constant value of W, the depth of penetration in the furnace using coke of particle size 41 mm is 125 cm, whereas the furnaces charged with coke between 51 and 56 mm in size have tuyere zones between 90 and 105 cm. The blast-furnaces at German works *M* are charged with coke at a particle size of 40–55 mm, while the larger furnaces at works *S* operate better at a size of 50–65 mm. Both the elliptical scatter bands for the individual readings and the mean

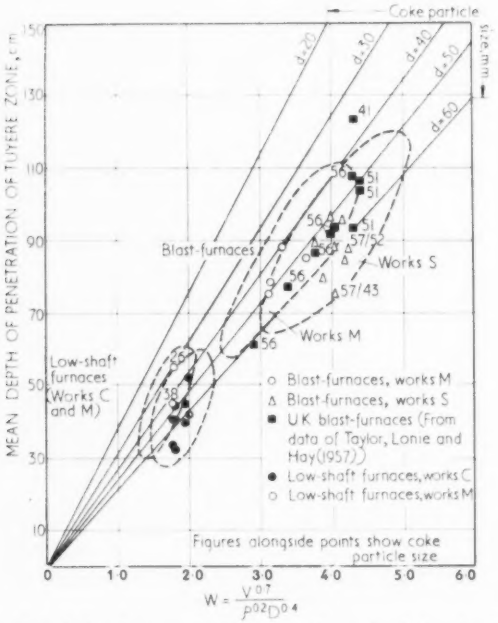
TABLE II Typical measurements of mean depth of penetration of the tuyere zone on blast-furnace I at works *M*

Tuyere no.	Depth of penetration of tuyere zone in cm	Remarks
1	75	Quantity of hot blast
2	106	43000 nm³/h
3	91	blast quantity per tuyere
4	89	6140 nm³/h
5	87	hot blast pressure 0·44 atm
6	69	hot blast temp. 360°C
7	71	tuyere dia. 180 mm

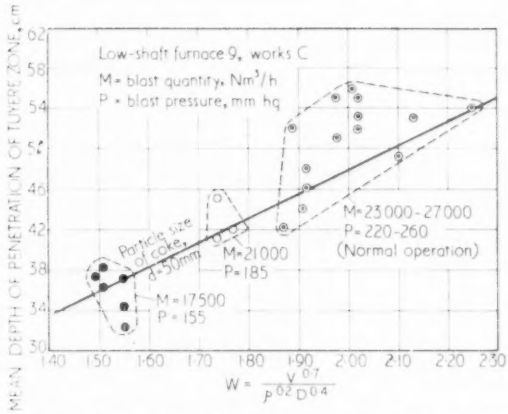
Mean depth of penetration of tuyere zone 84 cm



1 *Correlation between the pressure difference criterion and the mean depth of penetration of the tuyere zone of a blast-furnace*



2 *Correlation between the blast quantity and relative density, tuyere dia., and coke particle size and the depth of penetration of the tuyere zone*



3 *Depth of penetration of the tuyere zone as a function of the value of* W *with variations in the blast quantity and pressure*

values indicated by points for each furnace fit satisfactorily into the family of curves in Fig.2, thus confirming the general validity of equation (1) for blast-furnaces.

The low-shaft furnace at works *M* is usually driven on small coke at a size of 38 mm, and a corresponding tuyere zone of about 45 cm. Temporary changeover to a coke size of 26 mm led to an extension of the tuyere zone to around 54 cm.

The position of the scatter band for the low-shaft furnaces at works *C*, and the characteristic mean values for the individual furnaces, indicate a deviation from the predictions of equation (1), in the direction of reducing the depth of penetration of the tuyere zone. This reveals the special nature of the high temperature coke made from brown coal, as developed by Rammler and Bilkenroth[4]; this coke constitutes about 65% of the total fuel charged to this furnace. The mean particle size of the coke at the time of the tests was 45 mm; its packing density was about 1·4 times greater than that of coke made from hard coal, at the same particle size. The higher density of this high-temperature lignite coke produces a smaller depth of penetration than would be anticipated from its particle size.

In some of the blast-furnaces at works *S*, the particle size was determined on samples of metallurgical coke removed through the tuyeres during shutdown. With a constant particle size before charging, namely 57 cm, the coke in front of the tuyeres was between 41 and 55 mm in mean particle size. The test results can be subdivided into two groups, of mean particle size 41–46 and 50–55 mm respectively; at a constant value of *W* the corresponding depths of penetration were 73 and 86 cm respectively. If during its passage down the furnace the coke attains a relatively small particle size, say because of inadequate strength, the depth of penetration of the tuyere zone decreases, contrary to the change predicted by equation (1). If coarse particles of ore reach the tuyere zone they have a similar effect to coarse-particle coke. This gives a further reason for the relatively small depth of penetration of the tuyere zone in the low-shaft furnaces at works *C*, where at the time of the tests ore particles were continually reaching the tuyere zone, because of inadequate burden preparation.

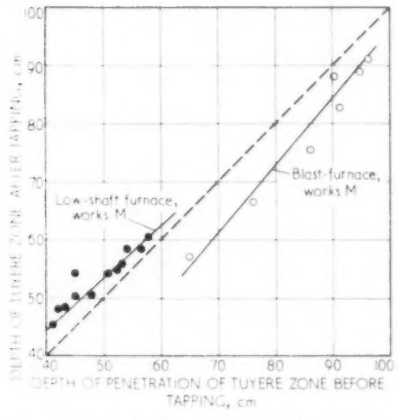
It can be stated that, provided actual conditions can be regarded as normal, equation (1) gives a satisfactory calculated value for the depth of penetration of the tuyere zone in low-shaft furnaces as well as blast-furnaces. We now wish to discuss further some of the special features occurring during the course of a campaign.

Discussion

It was possible to take measurements with varying quantities of hot blast, on low-shaft furnace IX at works *C*; the coke size varied slightly about the value of 50 mm. At blast volumes of 17500, 21000, and 23000–27000 Nm³/h and the corresponding pressures, the results obtained fall on both sizes of the straight line shown in Fig.3. It can therefore be stated that equation (1) remains valid even when the normal blast input is substantially reduced.

To check how far the accumulation of iron and slag in the hearth might influence the configuration of the tuyere zone, measurements were made on the blast-furnaces and low-shaft furnace at works *M*, immediately before and after an iron tap; the results are presented in Fig.4. According to these data, the depth of penetration of the tuyere zone in the low-shaft furnace is smaller before than after tapping; in the blast-furnaces the opposite relationship holds good. This finding is apparently attributable to simple differences in hearth design in the two types of furnace.

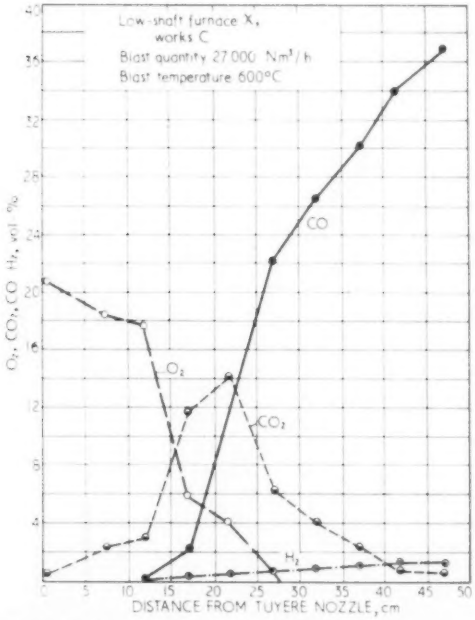
Thus, in *low-shaft furnace M* the centreline of the slag taphole is a mere 290 mm below the tuyere plane. Consequently, before tapping, the molten products in the hearth can relatively easily rise and reduce the extent of the tuyere zone. In the *blast-furnaces* on the other hand the distance between the slag taphole and the tuyere plane is 1200 mm, so that the accumulation


DEPTH OF TUYERE ZONE AFTER TAPPING, cm
DEPTH OF PENETRATION OF TUYERE ZONE BEFORE TAPPING, cm
Low-shaft furnace, works M
Blast-furnace, works M

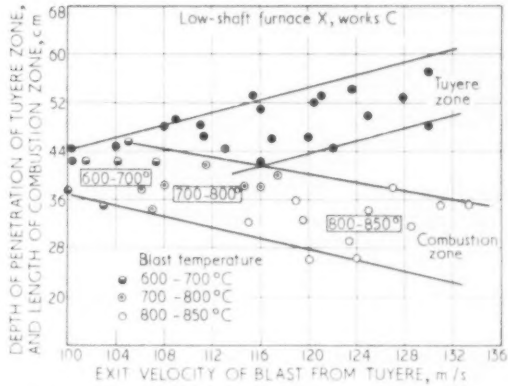
4 *Variation in the depth of penetration of the tuyere zone during the iron tap, under otherwise constant conditions*

of molten products should not influence the configuration of the tuyere zone. Moreover, as pointed out by Kinney[5] as long ago as 1929, the burden drops at an increased rate during the tap, so that this can disturb the tuyere zone.

Simultaneously with the investigations on the configuration of the tuyere zone, gas samples were taken, on some of the low-shaft furnaces at works *C*, using a water-cooled tube; the samples were taken at various distances from the tuyere nozzle. Figure 5, the first published graph of results of this type of series, is typical of a large number of similar investigations clearly indicating that the variation in composition of the gas phase on front of the tuyeres of a low-shaft furnace is basically the same as the well known variation in the blast-furnace.



5 *Variation in gas composition with increasing distance from the tuyere nozzle*



6 *Influence of exit velocity of the blast on depth of the tuyere zone and length of the combustion zone*

The next important question is to what extent the depth of penetration of the tuyere zone agrees with the extent of the zone within which the coke is burned. According to Bardin *et al.*[6] the limit of the combustion zone can be taken to coincide with the point where the gas phase contains 2%CO_2 or 30%CO, so that the following equation can be used for the length of the combustion zone:

$$L_m = \frac{L_{2\%CO_2} + L_{30\%CO}}{2} \quad \ldots\ldots(3)$$

In Fig.6, the depth of penetration of the tuyere zone and the length of the combustion zone have been plotted against the exit velocity of the hot blast at the tuyere. At blast temperatures of 600–700°C, the tuyere zone and combustion zone of the low-shaft furnace X concerned have about equal lengths. Increasing the blast temperature promotes faster reaction and thus shortens the combustion zone; on the other hand, if the quantity of blast remains constant, increasing its temperature extends the tuyere zone because of the increase in exit velocity. This finding cannot be applied generally, as it still needs support by investigations on other furnaces.

Finally, temperatures in front of the tuyeres were measured on the low-shaft furnaces in works *C* and the blast-furnaces in works *S*, using a partial radiation pyrometer. According to these findings, the temperatures arising in front of the tuyeres are at about the same level in both types of furnace, namely 1780°C. No correlation could be found either between the depth of penetration of the tuyere zone and the temperature readings, or between the thermal state of the hearth and the temperature in front of the tuyeres. This is indicated by the fact that no difference could be found in the temperatures in front of the tuyeres between two blast-furnaces, one making hematite pig with 2·50%Si and the other steelmaking pig with only 0·75%Si.

CONCLUSIONS

1. The depth of penetration of the tuyere zone is determined by the factors of temperature and pressure of the hot blast, tuyere diameter, and particle size of the coke. The formula derived by Taylor *et al.* is valid in general for both blast-furnaces and low-shaft furnaces.

2. There is a definite correlation between the depth of penetration of the tuyere zone and the quantity of blast entering from the individual tuyeres, as indicated by the pressure difference.

3. A small particle size of the coke favours an extension of the tuyere zone. If, however, the smallness of the coke is due to crushing as it passes down the furnace, this results in a reduction of the tuyere zone, associated with a deterioration in the state of the furnace. If the conditions are otherwise unvaried, while the iron is being tapped the depth of the tuyere zone increases in the low-shaft furnace; in the blast-furnace on the other hand it decreases.

4. The configuration of the combustion zone in front of the tuyeres in the low-shaft furnace follows the same laws as those long known in the case of the blast-furnace.

5. Between the extent of the combustion zone and the depth of penetration of the tuyere zone there are indications of a correlation which requires further investigation; at a constant blast quantity, increasing the blast temperature shortens the combustion zone.

6. No connexion could be found between the optical temperature measured through the tuyere peephole on the one hand and the depth of the tuyere zone or the thermal state of the hearth on the other.

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Acid smelting of foundry pig iron in low-shaft furnaces

Georg von Struve, Dr-Ing.

INTRODUCTION

THE IDEA OF SMELTING PIG IRON in a blast-furnace under a slag with a basicity factor of less than 1 is by no means new. As long ago as 1923 Brassert[1] showed that the operation of an acid slag led to the possibility of producing pig iron more economically. Nevertheless it was not until 11 years later that Paschke and Peetz[2,3] recommended after successful small-scale tests the use of the smelting process which had become known as 'acid smelting'. In the following years large-scale trials were undertaken in various German iron-works and in Corby (England) to learn the technological, metallurgical, and economic peculiarities of acid smelting.[4–14]

The main reason for developing this smelting process was the steadily worsening position of the iron ore market. It was becoming necessary to smelt in greater quantities the vast amounts of lean ores available in Germany and the UK. As these ores were low in iron content and mainly siliceous in character, the vast amount of slag required by normal basic slag working, and in consequence the coke consumption, would have risen to economically inadmissible levels. Because of the high cost involved it was impossible to consider beneficiation with accompanying agglomeration. A possible solution was seen in acid smelting in the blast-furnace provided there was no attempt to effect any large-scale removal of sulphur during the smelting process.

SYNOPSIS

This paper gives the results of two large-scale tests on acid smelting in low-shaft furnaces at the Calbe works, Saale, Germany. In these tests the slag ratio of about 1·12 was reduced to 0·9 and 0·7, a burden yield of 23% being maintained. Technological and metallurgical characteristics of acid smelting are described in detail, and the dependence of desulphurization and carburization on the slag ratio and the silicon reduction is demonstrated.

The principal conclusion of these tests is that the efficiency of pig iron production can be considerably improved by utilization of the acid smelting process. This may be attributed to increased economies in fuel and limestone as well as an appreciable increase in production. On the other hand the utilization of the acid smelting process results in a certain reduction of quality of the pig iron. This is characterized by an increasing sulphur content and a decreasing carbon content. But it is possible to improve the pig iron quality so that it is equivalent to that obtained by basic smelting. 1821

Note: Throughout this paper, 'tons' refer to metric tons.

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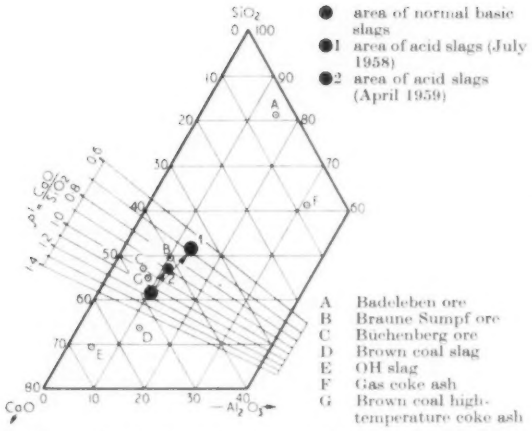


1 *State of the slags and the gangues of the charge materials in the* $CaO–SiO_2–Al_2O_3$ *system*

How much nearer are the prospects of using an acid slag in a low-shaft furnace developed for smelting inferior grade raw materials? As pig iron is being smelted on a production scale in low-shaft furnaces at Calbe (to date the biggest plant in the world doing this), and as the raw material position in central Germany is anything but favourable, it was understandable that the attempt should be made to link up with the experiments of the past to try to find new possibilities for improvement in the economics of smelting.

This paper describes the trials undertaken in Calbe by the Eisenhütten-Institut der Bergakademie Freiberg under the direction of the author since the middle of 1958. On the basis of the satisfactory results obtained, several furnaces have been operating with an acid slag since March 1959. It appears that acid smelting may once more become significant for conventional blast-furnaces, since five out of six blast-furnaces at Hüttenwerk Watenstedt Salzgitter AG were converted to acid smelting some time ago.[15]

ACID SMELTING IN THE BLAST-FURNACE AND THE LOW-SHAFT FURNACE

There were some differences in the method of working between the smelting trials done previously in the blast-furnace and those made by the author at Calbe. Lennings[6, 7] reported on the production of Thomas pig iron in three blast-furnaces at the Gutehoffnungshütte in Oberhausen. It was proved pointless to distribute equally the lean German Dogger ores, whose smelting properties were being assessed, in all the blast-furnaces with normal basic slag practice. For that reason, in October 1937, the German ore was smelted in a blast-furnace, together with rich siliceous imported ore, by the acid process (Fe content from the German ore: 48·1%), while the high-grade ore was left to the other two furnaces with basic slag operation. In November the amount of German ore in the acid furnace was raised to 100%. In both cases a CaO/SiO_2 slag ratio of 0·8 was aimed at. Desulphurization of the acid-smelted iron took place outside the blast-furnace in a ladle with a mixture of granulated calcined soda and limestone splinter. Economic considerations which were drawn into the assessment of the usefulness of the acid smelting process are based on a comparison of the operating results from all three furnaces before and during the trials.

TABLE I Data of acid and basic smelting process on furnace 9

	June 1959	5–9 July 1959
Basicity $p' = CaO/SiO_2$	1·12	0·69
Burden yield, %	23·3	23·5
Pig-iron production, tons/24 h*	102·8	107·1
Pig-iron production, tons/h†	4·30	4·64
Coke consumption, kg/ton pig iron	1890	1627
Hearth load, $\frac{\text{kg coke}}{\text{m}^2\text{h}}$	750	674
% of BHT-coke	54·3	77·7
Volume of blast, Nm^3/h	26–27000	24–26000
Blast temp., °C	800–850	800–850
Proportion of scrap and runner iron, kg/ton pig iron	359	314
Limestone, kg/ton pig iron	916	87
Slag volume, kg/ton pig iron‡	1988	1990
Composition, %		
CaO	45·14	32·14
SiO_2	40·35	46·85
MgO	2·54	2·34
Al_2O_3	10·05	11·24
Fe	0·78	2·71
Mn	0·51	1·92
S	0·69	0·57
(S)/[S]	8·3	2·9
Pig-iron composition, %		
C	3·57	2·69
Si	3·13	3·54
Mn	0·81	0·68
P	0·63	0·61
S	0·083	0·200
Manganese charge, kg/ton pig iron	22·9	47·3
Manganese yield, %§	38·3	15·0
Iron yield, %§	96·2	89·2
Occurrence of 'puddled' iron, kg/ton pig iron	15·1	16·1
Low-shaft furnace gas analysis, vol-%		
CO_2	5·4	5·8
CO	36·4	35·2
H_2	1·7	1·7
CH_4	0·2	0·2
Indirect reduction, %	36·9	36·2

* Actual production in the furnace
† Production related to operating hours
‡ Theoretically calculated
§ Nett, after deducting dust losses

Hahnel worked differently in 1938 in the production of Thomas pig iron in a blast-furnace at the Maxhütte Works in Unterwellenborn.[9] By increasing the ore input and simultaneously reducing the limestone additions he lowered the basicity of the slag step by step from 1·59 to 0·22 (ultra acid slag). The trials were over a period of 2½ months so that there was occasionally a guarantee of reaching a stability in the operating conditions. It is characteristic that the burden yield rose progressively from 24·9% to 38·5%. Further trials ascertained the burden mixtures which showed the lowest costs at a slag figure of 0·3. At this point the burden yield was 11·2% or 11·5% higher than in the usual smelting process with basic slag. With slightly acid slags the pig iron, which contained sulphur, was treated with soda in the ladle; with very acid slags it was re-cast into the hearth of a basic working blast-furnace and was also desulphurized with soda after the blast-furnace.

At Hoesch AG, Dortmund, in June and July 1938, Schumacher[10] finally carried out a full-scale trial with Salzgitter ore. He compared the effect of the acid smelting of this ore (CaO/SiO_2–0·75) with that of an additional preparation of the burden (sintering of the ores and smelting of burnt lime) and, theoretically, with the basic smelting operation without preparation of the burden (CaO/SiO_2–1·4).

These cases had in common the fact that the burden

yield changed from trial to trial, and that either the basic smelting of the ore reviewed was not investigated at all or else there were changes of principle in the method of working.

As opposed to this, the author's trials on low-shaft furnace no.9 at Calbe from 1 to 9 July 1958 started from the normal basic smelting of the home ores and reduced only the CaO/SiO_2 slag ratio from about 1·10/1·15 to 0·70. In this way, therefore, there were no new principles introduced. Finally, it is worthwhile noting that an attempt was made to keep the burden yield constant. This seemed to be desirable for the purpose, so that especially favourable results from the trial would not be invalidated in advance by the relatively high scrap input of 320–330 kg/ton pig iron. The proportion of home ores was therefore increased on the average from 2·79 to 3·64 tons/ton pig iron; the scrap input was slightly reduced. The experiments were to confirm to what extent the operating data, the quality of the pig iron, and the costs per ton of pig iron change with acid smelting. The investigation of a special process for the desulphurization of the iron after smelting in the low-shaft furnace was not taken into consideration, since during the period of the trial all the iron went through the 300-ton mixer. The month of June was used as a comparison for the results.

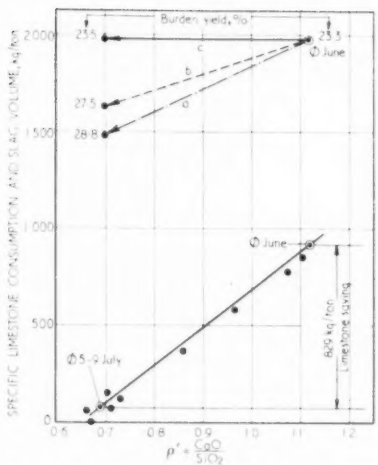
Since April 1959 four out of eight low-shaft furnaces at Calbe have been working with a slag ratio of 0·9 under otherwise similar conditions as obtained in the trial run of furnace no.9 in July 1958. The first 20 days of April are therefore also used as a yardstick of comparison.

DATA AND OPERATION OF THE LOW-SHAFT FURNACES

Within the framework of a summary it is not possible to go into all the details of the trial run and of the present operation of the low-shaft furnaces in Calbe. Attention, therefore, is directed to several essential features.

There have already been several reports[16–18] about the structural and metallurgical peculiarities of the low-shaft furnaces in Calbe under normal working conditions. The state of the gangue of the most important raw materials in the three materials system, lime–silica–alumina, is shown in Fig.1. The magnesia contents are occasionally below 5% (6% in OH slag) so that they can be disregarded in this account. The proportion of the very siliceous Badeleben ore amounts to about one-third of the ore burden. The slightly basic ash of the brown-coal high-temperature coke is notable.[19] The state of the normal low-shaft furnace slag in Calbe was entered on Fig.1. The magnesia content of this slag varies between 2·5% and 3%. Field 1 gives the slag point of the trial series in July 1958 with a CaO/SiO_2 ratio of about 0·7. Field 2 is the area of acid slags of furnace series 1 in April with a ratio of about 0·9.

Starting from the slag figure of 1·19, which corresponded to the daily average at 30.6.58, the basicity of the slag was reduced by 0·1 each day, so that at the beginning of 5 July the figure of 0·7 was reached. Despite large variations in the analysis of the charged materials, this figure remained almost constant until 9 July. After that the series of trials had to be discontinued because of unforeseen operational circumstances so that only five days could be used for the evaluation of true acid working.

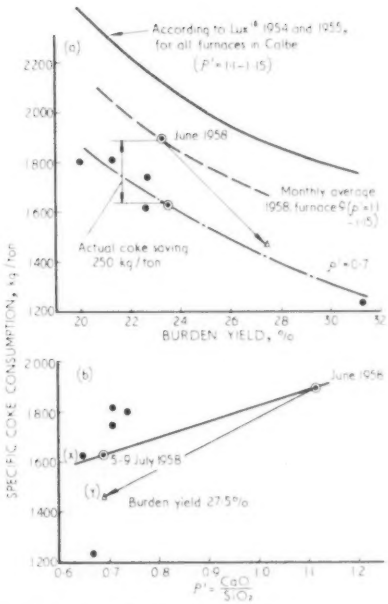


Calculated slag volume; *a*, with reduction of limestone and coke charge; *b*, as *a* but taking into consideration a greater loss of iron and manganese into the slag; *c*, trial series, 5–9 July 1958

2 *Limestone consumption and slag volume with various slag ratios*

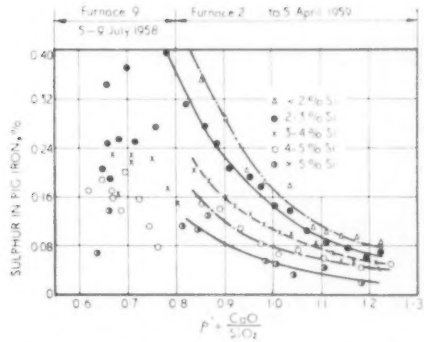
Taking into consideration the short burden throughput time of only 2·7–3 h, it is not very far wrong, however, to count these five days as being equivalent to a 12–14 day blast-furnace trial. If a delay of 3–4 h between the beginning of the running and the tapping slags is taken into consideration, then there are only slight differences in the basicity. In individual cases the tapping slag was somewhat more basic. For this reason, and on account of a several times greater amount of running slag, the analyses of this are taken as a basis for further evaluation.

In Table I the most important operational figures for the basic and acid working of the furnace is compared. The burden yield remained unchanged as intended. The rise in production of 0·34 tons/operating h, because of the acid process, is therefore interesting. This corresponds to an increase in capacity of 7·9% and more notable, since the entry of iron and manganese into the slag was increased and the hearth load dropped, both of which might have worked against an increase in production. The reason for the lower coke throughput is the throttling back of the blast volume by about 2000 Nm^3/h (7·7%) because the charging could not always keep pace with the increased top throughput due to the lower charge weights and because it was not desired to allow the charging column to fall below 4·8 m (as measured from tuyere level). Driving the furnace full out, an increase in capacity of about 15% could be expected. This is due to the quicker movement of the burden through the furnace and the more even working of the furnace as a result of changed slag properties as well as a greater ore/coke ratio. It is worth mentioning finally that the amount of brown-coal high-temperature coke (BHT-coke) increased from 54·3 to 77·7%. The occurrence of 'puddled iron', a form of iron in small droplets removed with the running slag, changed only slightly from 15·1 to 16·1 kg/ton of pig iron. The somewhat hotter working of the furnace should, however, be


According to Lux[18] 1954 and 1955, for all furnaces in Calbe
(P' = 1·1–1·15)
June 1958
Monthly average 1958 furnace 9 (P' = 1·1–1·15)
Actual coke saving 250 kg/ton
P' = 0·7
SPECIFIC COKE CONSUMPTION, kg/ton
BURDEN YIELD, %
June 1958
5–9 July 1958
Burden yield 27·5%
P' = CaO/SiO2

(X) actual results at 23·5% burden yield
(Y) theoretical results for reduction of limestone consumption and slag volume

3 *Specific coke consumption with acid smelting in the low-shaft furnace at Calbe*



4 *Influence of the slag ratio on the sulphur content of the pig iron with differing silicon contents*

taken into consideration as this might have increased the amount of 'puddled iron' arising.

The specific limestone usage for each day of the trial is entered over the CaO/SiO_2 ratio in Fig.2. The linear relationship is clearly seen if the average value for June is taken as a basis of comparison. It shows that a saving in limestone of 829 kg/ton pig iron is obtained with the burden condition of that date by altering the CaO/SiO_2 ratio from 1·12 to 0·69. It also confirms the calculation made before the trial that a slag ratio of about 0·65 is the lower limit for the acid smelting process at Calbe, unless the amount of very acid constituents in the burden is raised.

From Fig.2 it may be deduced that the specific amount of slag, with the exclusive reduction in the limestone and coke necessary for smelting, must theoretically fall from 1988 to 1485 kg/ton of iron, since the burden yield then rises from 23·3% to 28·8%. The slag volume was calculated on a silica balance taking dust losses into account. If the higher loss of iron and manganese into the slag is taken into account, then a reduction in slag volume of 375 kg/ton could still be expected. In fact, however, the burden yield was held almost constant, so that a reduction in the specific slag volume was not obtained.

Figure 3 gives a picture of the fall in coke consumption. Theoretically a coke saving of 436 kg/ton should be achieved according to a simple heat balance, if the previously illustrated change of limestone and coke volume are taken as a basis for comparison. This interdependence, however, is not decisive in this series of trials because of the increasing burden yield. To come to a real assessment of the possible coke saving through acid smelting, the trial results were compared with the burden yield. In spite of the small number of points there is quite a definite relationship, with which the theoretical value (triangle) closely coincides. If the curve for furnace 9 (using basic slag), built up from the 1958 monthly average figures, is added to the graph, then the possible coke saving for any desired burden yield can be read off. For a burden yield of 23·3%, this saving amounts to 250 kg/ton of pig iron. A comparison with the conditions obtaining for all the low-shaft furnaces at Calbe between 1954 and 1955 according to Lux[18] shows how the operating conditions have improved over the past four years. It must be noted, however, that furnace 9 was the furnace with the lowest specific coke consumption.

It might be of interest in connexion with limestone and coke saving to turn to a consideration of current operating data during the first 20 days of April 1959 (slag ratio, 0·91). As a basis for reference the monthly averages between December 1958 and February 1959 were used, as by March several furnaces had been switched over experimentally to smelting with a slightly acid slag. With an increased burden yield of 0·9%, a limestone saving of 240 kg/ton on the average and a fall of about 90 kg/ton in the specific coke consumption was achieved. This result was somewhat less favourable than in the trials of July 1958. This was because at that time about 600 kg/ton of pig iron of self-fluxing easily smeltable ores were used in the burden, whereas in April 1959 the specific consumption of the siliceous Badeleben ore was increased. Besides the analysis of the charge materials in the comparative months shows considerable differences. Despite this, savings in limestone and coke are considerable even through just slightly acid smelting.

Hahnel[9] observed that, if the slag ratio in acid smelting in the blast-furnace dropped below even 0·82, there was such an increase in the slag viscosity that difficulties arose in getting the slag to flow. By adding to the burden some soda desulphurization slag, which is characteristically high in alkali content, this could be counteracted. The blast-furnace slag remained so fluid that basicities of as low as 0·22 were reached.

In the author's trials in furnace 9 a certain increase in viscosity was also observed when the basicity fell below 0·75; however, the slag could be tapped without difficulties if the furnaceman worked carefully. Soda desulphurization slag which had been held ready in case of emergency, was therefore not required in the burden. The fact that the slag remained sufficiently fluid must be attributed to the high FeO content, and

above all to the excessively high MnO content due to a bad manganese yield and an increased manganese input to ensure the maintenance of the prescribed manganese content of the pig iron. The slag pancake was characterized by a light to dark blue colour at times verging to green.

During the trials it was often possible to observe that, when the furnaceman was careless, the slag began to flow over the runners very shortly beyond the Lurmann-type slag tuyeres. Here the dependance of the viscosity of the slag on temperature was clearly noticeable. As a result of the cooling on the side walls, and particularly on the surface, more or less strong hard crusts were formed, which tended to impede the flow of the slag. It was proposed, therefore, to reduce the loss of heat upwards by the covering of the runners or, better still, by additional heat.

Disruption in furnace operation, any tendency towards hanging on the part of the charge, or uneven gas distribution throughout the furnace were not in evidence either in furnace 9 in July 1958 or in furnace series 1 in April 1959.

REDUCING CONDITIONS AND PIG IRON QUALITY

The average composition of the pig iron and slag is given in Table I. The pig iron produced in the acid process with its 0·41% higher silicon content is notable. Even if it is conceded that the furnace was run without any risk, i.e. somewhat hotter than usual, then it can still be said with safety that a more extensive reduction of silicon follows in acid smelting owing to the higher silica availability in the burden and the greater SiO_2 activity. This is particularly pronounced under the conditions then obtaining in Calbe, i.e. smelting of the very siliceous Badeleben ore in lumps up to 150 mm and, presumably, also the very high scrap charge. Charging agglomerated fines, if possible in the form of self-fluxing pellets, might lead to improvement with regard to a lower silicon reduction. But even with the present quality of the burden, the silicon content of the iron could be kept to the normal limits by a somewhat colder method of working without suffering from a worse flow of slag. Furnacemen should not be misled by the often dark colour of the slag in assessing the heat condition in the hearth after the slag sample.

Despite the fact that the manganese input was more than doubled, the manganese content of the iron fell somewhat. It is known that the reduction of manganese declines as the slag becomes more acid. A manganese yield of 15% is exceptionally bad and no doubt the high slag bulk of about 2 tons/ton of pig iron is to blame for this. Despite this, and contrary to published views, the author believes that increased use of manganiferous materials is worthwhile in every case, especially where a cheap manganese-bearing raw material like ferromanganese slag is concerned. The higher MnO content has such a beneficial effect on slag viscosity that it makes the addition of any other fluxing compound such as alkali oxide or magnesia unnecessary.

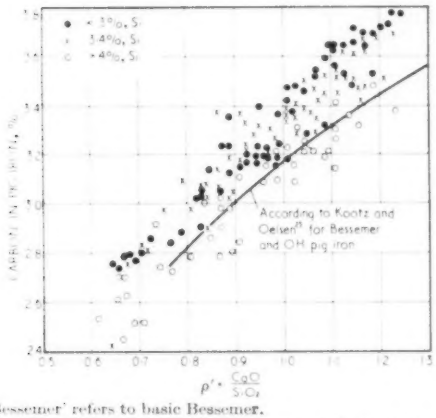
In acid smelting, the higher entry of iron into the slag, characterized here by the iron yield, should be noted. This is the cause of the greater amount of material charged, which must be taken into account in economic considerations. In Table I the net yield is shown, i.e. deducting dust losses.

Before discussing the change in the carbon and sulphur contents of the iron, due to acid smelting, it is necessary to answer a question which is repeatedly asked. How great is the degree of indirect reduction if the large amounts of limestone are left out of the burden? As Table I shows, the degree of indirect reduction remains almost constant. This result is first of all rather disconcerting and seems to contradict blast-furnace practice, according to which, for example, the replacement of limestone by burnt lime causes an increase in indirect reduction because of the smaller concentration of CO_2 in the stack gas. In this case the indirect reduction was calculated according to a formula in which regard was paid separately to the dissociation of the carbon dioxide from ores and limestone corresponding to the degree of reduction then obtaining.[20] Other formulae, e.g. according to Bansen[21] or Ramm,[22] which do not recognize this difference, would indeed indicate an improvement in indirect reduction in acid smelting which the rising CO_2/CO ratio would also point to.

In practice, however, almost 70% of the limestone CO_2 is dissociated owing to the short throughput time in the low-shaft furnace. This gives rise to CO with considerable coke consumption, which helps in the reduction. It is therefore clear that with the lessening of limestone input the reducing power of the gas will decline rather than improve. Finally it is necessary to take into account the fact that the proportion of self-fluxing ores with CO_2 contents up to 20% used in acid smelting was increased, which presumably counterbalances to some extent the more favourable reducing effect of the shaft gases due to the lower hearth load.

Concerning sulphur and carbon contents of acid and basic produced pig iron, all taps from furnace 9 between 5 and 9 July 1958, and those from furnaces 2–5 in the first 20 days of April 1959, were evaluated. In the second case, however, only the statistically determined averages were discussed. Figure 4 shows, for the current operation down to a slag ratio of 0·8, a definite increase in the sulphur content of the iron with decreasing slag basicity, from which poor desulphurization effect of acid slags results. If the values are grouped according to the silicon content of the iron, it becomes apparent that an increasing silicon reduction brings with it an improvement in desulphurization. These casual relationships mean that the laws discovered in the laboratory by Oelsen and Maetz[23, 24] and confirmed by other authors in assessing the operation of the blast-furnace, also hold good for the low-shaft furnace. Finally, Fig.4 shows that the influence of slag basicity outweighs that of the silicon reduction. Therefore, it appears economically untenable to strive after better desulphurization of the iron by hotter working in the low-shaft furnace. It is much more effective and much cheaper to desulphurize the iron outside the smelting unit. With desired silicon contents of under 3%, the sulphur content of the iron produced in Calbe under present normal conditions will hardly ever exceed 0·4%.

It is of interest that the sulphur content of the pig iron during the trial period on furnace 9 did not coincide with the curves shown on Fig.4 but took up a more favourable position on the graph. This may be explained on the one hand by the lower sulphur load of 18 kg/ton as opposed to 25 kg S/ton of pig iron, and on the other hand by the 1·5% higher MnO content of the slag. A gradation of the sulphur values according to

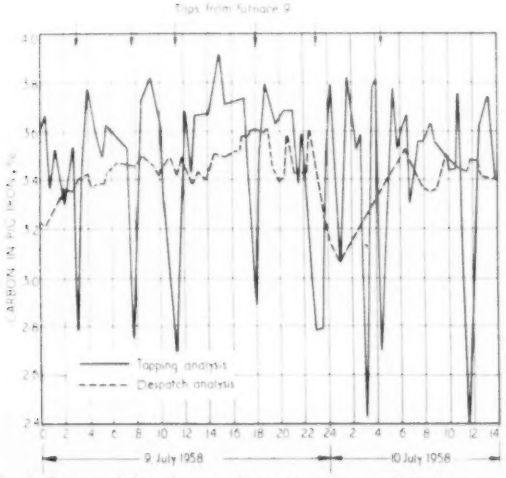


'Bessemer' refers to basic Bessemer.

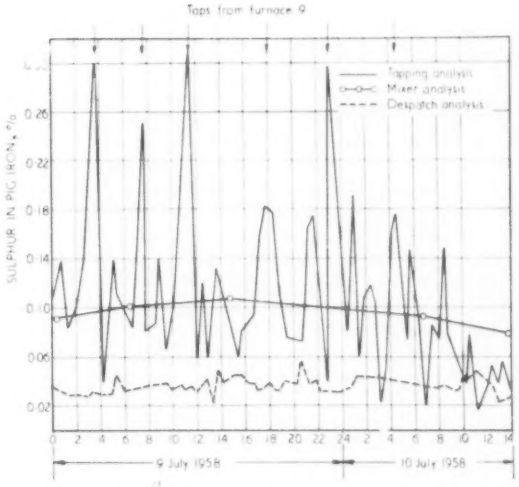
5 *Relationship between the carbon content of the pig iron and the slag ratio with differing silicon contents*

the silicon content is similarly obvious as in furnaces 2–5 with higher slag ratios.

While the increase in sulphur content in the pig iron with acid smelting is acceptable, when a suitable desulphurization process is added after the iron has been smelted then the fall in carbon content in the production of foundry pig iron must be considered a disadvantage. Figure 5 shows a clear relationship between the carbon content of the pig iron and the slag basicity for all evaluated taps. According to this, a drop in carbon content of 0·2% corresponds to a fall in the basicity of 0·1. It is not yet possible, on the basis of operating results, to give a satisfactory explanation for this. Undoubtedly the reducing conditions, which are decisive for the carbonization of the pig-iron in the bosh and hearth, are worsened by the viscous slag which is presumably cooler owing to the lower formation temperature. Experiments carried out by Wenzel[26] show that carbonization conditions in the shaft and bosh of the furnace do not deteriorate in the acid smelting process. The majority of the points in Fig.5 lie above the line found by Kootz and Oelsen for basic Bessemer and OH pig iron produced by blast-furnaces, because of the hotter working of the low-shaft furnaces. The taps with higher silicon content



6 *Influence of the mixer on the carbon content of the iron*



7 *Influence of the mixer on the sulphur content of the iron*

usually show a lower carbon content, which is explained by a downward movement of the saturation limit.

Finally, the furnace operator is mainly interested in making a saleable pig iron using the acid slag process. If only one of eight low-shaft furnaces is worked with an acid slag, as was the case in the July trials, and if the mixer is used, then there is no need for special arrangements for the acid-smelting pig iron. Figures 6 and 7 show the equalizing effect of the mixer on the iron analyses for any desired period of time. Although carbon contents of 2·7% and under were frequently received from the furnace, only once, at 1 pm on 10 July, did the carbon content of the pig iron, for further despatch, fall below 3·1%, since at that time three consecutive taps with exceptionally low carbon contents occurred. As far as the sulphur content was concerned, there was absolutely no indication of any effect on the despatch analyses owing to the acid-smelted iron. This is due on one hand to the desulphurizing effect of the mixer, and on the other hand all ladles from the mixer were desulphurized by the addition of three bags of soda as in the normal basic smelting operation. This corresponds to a soda consumption of 7–8 kg/ton of pig iron.

During April 1959, when four low-shaft furnaces worked with a slag ratio of 0·91 and four others worked with a ratio of 1·08, the following average analysis for foundry pig iron could be achieved:

2·80%Si 0·68%Mn 0·60%P 0·033%S 3·34%C

From the quality point of view there were no doubts about continuing this method of smelting. If more furnaces should be switched over to working with a slag basicity of 0·7, then it would be impossible to do without additional desulphurization and carbonization after smelting.

Further details in this respect will be published later. It is sufficient to say that already in individual full-scale trials developed from laboratory experience, a very satisfactory degree of desulphurization is shown by the injection of Buna lime (a powdery hydrate of lime containing 63–65%CaO which occurs in the production of carbide) into a ladle by means of compressed air or nitrogen through a graphite lance. The use of

powdered calcium carbide as a desulphurizing material may be recommended to achieve a simultaneous carburization of the pig iron.

ECONOMIC CONSIDERATIONS

In all trials intended to introduce a new smelting process into practice, economic considerations cannot be overlooked. For the time being such considerations can only be approximate as it is not certain whether the savings achieved can be obtained in continuous operation with furnaces working with a slag ratio of 0·7. In addition, only current reckoned prices for the work at Calbe are known, and these do not reflect properly the true value of the raw materials.

If the method of working during the series of trials on furnace 9, i.e. with constant burden yield, is taken as a comparison, beginning with the costs of July 1958, a false conclusion would be reached. It would be better to reckon back beginning with the acid smelting and to determine the cost per ton of basic pig iron with increased use of home ores. The difference would then give the possible reduction in costs.

To simplify, it is sufficient for a rough calculation to estimate the cost reduction where the burden yield rises to 27·5%, taking into consideration the greater entry of iron and manganese into the slag. A reduction of 700 kg limestone and 400 kg coke/ton of pig iron, where it is accepted that the BHT-coke proportion amounts to 50%, would then mean a lowering of the raw material costs by 34·03DM per ton of pig iron. Against this there is the extra consumption of 150 kg/ton of pig iron of home ore and 170 kg of ferromanganese slag/ton of pig iron which raises the raw materials cost by 4·16DM/ton of pig iron.

If in the first instance one disregards a possible alteration of the credits, but takes into account on the other hand a 7DM/ton reduction of working and overhead costs in a possible production increase of 10%, then that would give a cost reduction of 36·87DM/ton of pig iron. It would not be wrong to subtract a maximum of 7DM/ton of pig iron for the additional desulphurization and carburization. The final result would, therefore, still give a saving of 29·87DM, i.e. about 30DM/ton of pig iron. This result of practical trials fits in very well with the value given by Tischendorf based on theoretical considerations for the Calbe Works using the acid smelting process.[27]

When one speaks of the cost build-up and the cost development in acid smelting, it must not be forgotten that a very difficult problem must be solved quickly, i.e. the subsequent treatment of the acid slag. It is not just a question of looking for a suitable use for the slag, for there are clear ideas on this. For the German Democratic Government there is, however, the important question: what basic and equally cheap raw material could be supplied in future to part of the cement industry if acid smelting is introduced on a large industrial scale?

CONCLUSION

As the low-shaft furnace is a suitable smelting unit for low-grade raw materials, it is very suitable for the use of acid smelting. An analysis of two large-scale production trials in the low-shaft furnace plant in Calbe where in one case the slag basicity was 0·7 and in the other 0·9, is made, and both the technological and metallurgical peculiarities of acid smelting are fully discussed. It can be confirmed that a change in the technical data, the quality of the pig iron, and the economics can be reckoned with in the low-shaft furnace as in the blast-furnace. All in all, the experiments to date have been so successful that no doubts are raised about the introduction of acid smelting in continuous operation. In this way pig-iron production may receive a new and encouraging impetus, especially in countries which do not have sufficient basic raw materials at their disposal.

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Changes in mechanical properties during the fourth-stage tempering of a chromium steel

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SYNOPSIS

True-stress true-strain data have been measured as a function of time and temperature during the isothermal tempering of a hardened, 11·4% chromium steel. At each tempering temperature the strain-hardening index and the 1% proof stress show a dependence on tempering time which is not parallel to that revealed by hardness measurements. The times at which the former properties change are correlated with microstructural changes.

Some evidence is presented to suggest that Cr_7C_3 is formed by a two-stage process. 1812

EXPERIMENT

THE VARIATION of hardness with time at temperature is an accepted method of following the changes which occur during the tempering of alloy steels. The formation of an alloy carbide is generally accompanied by an increase in hardness which passes through a maximum when the alloy begins to over-age. However, in straight-chromium steels the as-quenched hardness remains almost constant until the over-ageing stage is reached, yet it is known from metallographic studies that during this time a sequence of carbide precipitations occur. The purpose of the investigation was to examine the variation with isothermal tempering time of other mechanical properties, and to determine if any of these properties are more sensitive than hardness to the structural changes which occur during the tempering of chromium steels.

A chromium steel was used containing 0·12%C, 0·57%Mn, 0·2%Si, and 11·40%Cr. This was received as $\frac{1}{2}$-in. dia. rolled bar from which tensometer specimens were prepared after a suitable homogenizing anneal. The test pieces were austenitized in vacuum at 1100°C for 30 min and quenched directly into an oil-bath. Isothermal tempering treatments for up to 100 h were given in the temperature range 400–700°C and the treated specimens were stressed to fracture at a strain rate of 0·01 in./in. min in a Hounsfield tensometer. The load, the extension, and the minimum diameter were measured continuously. The true-stress true-strain plot had a slight curvature in the strain-hardening region so that the index could not be measured without some uncertainty. However, the values are considered to be accurate to within 3%.

Dilatometer specimens were austenitized and quenched by the same procedure and volume changes during isothermal tempering were measured with a

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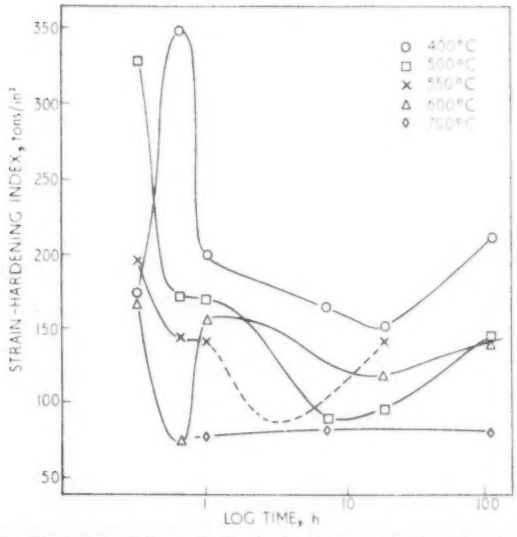
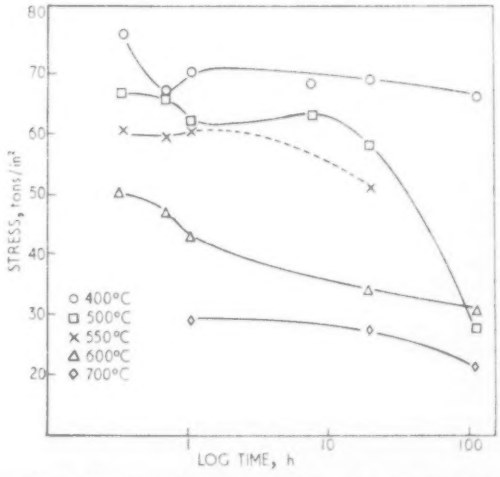
The authors were with the Department of Metallurgy, The University, Liverpool, when this paper was written.



1 *Variation of the strain-hardening index with time at various tempering temperatures*



2 *1% proof stress v. time at various tempering temperatures*

dilatometer which was sensitive to a change of 0·04% in the specimen length.

RESULTS

The variation of the strain-hardening index with time at temperature is shown in Fig.1 and the corresponding change in the 1·0% proof stress is shown in Fig.2. Generally it is found that at constant temperature the proof stress increases with time when the strain-hardening index decreases and vice versa. However, the magnitude of the variation of the index is considerably greater than the corresponding change in the proof stress, and the structural changes are more readily correlated with the former property.

During tempering at 400°C the strain-hardening index increases rapidly to a maximum and then falls sharply. The initial rise was not obtained at higher tempering temperatures, but it is possible that the maximum was already exceeded at the shortest tempering times (20 min) which were used. It was not considered advisable to employ shorter tempering times because a finite time was required for the specimens to reach thermal equilibrium.

Following the rapid fall, the curves of index *v.* time at 400, 500, and 550°C exhibit an abrupt change of slope, followed by a minimum. The location of the minima was not determined precisely, but they appear to move to shorter tempering times and to lower values of the index as the temperature is increased. No results are available for times between 1 and 17 h at 550°C. The broken line in Fig.1 is drawn in relation to the curves at other temperatures to indicate the probable location of the minimum. At 600°C the first minimum in the curve is followed by a second, more shallow, minimum after about 20 h at temperature. After 1 h at 700°C no further change occurs in the strain-hardening index, which suggests that equilibrium is obtained. With shorter times at this temperature the results showed considerable scatter.

Some austenite may be retained when the steel is quenched to room temperature, and the breakdown of this austenite during tempering could influence the changes in the mechanical properties. The amount of austenite retained in the present tests is not known precisely. Attempts to detect the presence of austenite by X-ray analysis were unsuccessful. The rates of volume change during the first and third stages of tempering were essentially the same in quenched and in refrigerated specimens. Therefore, it was concluded that the amount of austenite retained in the normal quench was negligible and was unlikely to affect the test results.

By comparison of the mechanical properties with the microstructures, it is seen that the changes in slope of the strain-hardening index *v.* time and the proof stress *v.* time curves are related to the structural changes which occur during the tempering process. Table I summarizes the changes in the structure and in the strain hardening parameter after tempering at 500°C. A similar correlation was obtained at the other tempering temperatures which were examined.

TABLE I Correlation of changes observed after tempering at 500°C

Time, h	Microstructure	Strain-hardening index, tons/in²
0·5	Fe_3C present	190 rapid fall
1·0	Fe_3C dissolving Precipitation of Cr_7C_3	168 steady fall
4·0	Fe_3C dissolving General precipitation of Cr_7C_3	116 steady fall
8·0	Precipitation and growth of Cr_7C_3	100 rough location of minimum
50·0	Fe_3C completely dissolved Coarsening of Cr_7C_3 in matrix First signs of $M_{23}C_6$ at grain boundaries	122 steady rise

DISCUSSION

At 400°C the initial rise in the index and the fall in the proof stress is most probably due to the breakdown of the martensite and the resultant increase in the plasticity of the material. The minimum in the index corresponds to the start of over-ageing of Cr_7C_3. A chromium carbide is first detected microscopically during the initial rapid fall of the index. Thus, it appears that the formation of the alloy carbide increases the resistance to plastic deformation up to the time when over-ageing occurs. The second minimum in the index, which was obtained after about 20 h at 600°C, is probably due to the formation of the massive carbide $M_{23}C_6$. This carbide is first detected microscopically after shorter times at temperature, but it is possible that, as in other ageing mechanisms, there is a temporary balance between two processes which overlap on the time scale. Electron microphotographs of specimens tempered for between 8 and 50 h at 500°C showed that Cr_7C_3 particles are agglomerating in some areas of the matrix while in other areas Cr_7C_3 is going back into solution and $M_{23}C_6$ is forming.

Most of the features of the strain hardening index *v.* time curves are thus readily interpreted in terms of the accompanying changes in the microstructure. However, the change in slope of the curve which is obtained after tempering for about 1 h at 400°C or for 40 min at 500 or 550°C cannot be explained in this way. Dilatometric studies showed that the third stage of tempering is completed in shorter times at each of these temperatures, and metallographic examination indicates only the formation of a chromium carbide with accompanying re-solution of the cementite. The evidence for the change in slope is based upon the location of only three points on each curve, but the fact that the points are reproducible and similar behaviour was obtained at three tempering temperatures indicates that the observation is not a result of experimental errors.

A possible interpretation is that the first chromium carbide Cr_7C_3 is formed by a two-stage process, the first stage of which may be either the clustering of chromium and carbon or the formation of a transition carbide. A carbide with carbon content similar to that in Cr_7C_3, but with a NaCl type lattice, was identified by Goldschmidt[1] in high-carbon high-chromium steels. This constituent was also reported by Lashko and Nesterova[2], but Kuo[3] suggested that the constituent was austenite which was undissolved during extraction of the carbides. A carbide similar to that reported by Goldschmidt and with only transient existence may have formed in the present study, transforming rapidly into the trigonal carbide after only a short time at temperature. Coutsouradis and Habraken[4] have drawn attention to the fact that the X-ray pattern, established by Westgren and generally accepted for

Cr_7C_3 in 0·2% carbon steels, differs from that found for the carbide extracted from steels of higher carbon content. Thus it is clear that the structure of the first-formed chromium carbide is not uniquely established for all compositions of the parent alloy and the possibility cannot be discounted of the existence of a transition carbide phase before the formation of the stable carbide.

CONCLUSIONS

The strain-hardening index is more sensitive than hardness to the structural changes accompanying the later stages of tempering in chromium steels. In the temperature range 400–700°C there is a good correlation between the changes with time at temperature of the strain-hardening parameter and the changes in the microstructure. At constant temperature the rate of change of the index is altered markedly during the time interval when the first chromium carbide is separating. This change occurs before the over-ageing of the carbide is detected microscopically and is interpreted as evidence of a two-stage process in the formation of the chromium carbide Cr_7C_3.

ACKNOWLEDGMENTS

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Effects of diffusion on corrosion of metals by fused salts

C. Edeleanu, J. G. Gibson and J. E. Meredith

SYNOPSIS

On considering the analogy between the corrosion of alloys in fused salts and in aqueous systems, the physical differences of pressure, temperature, etc. may be expected to have a profound effect on the chemical processes involved. In equimolar NaCl, KCl *at 800°C it is almost invariably found that stainless steels form internal voids. These are considered to be the result of internal diffusion and selective oxidation of one or more of the components. Metallographic examination of the corroded alloys indicated in this instance that the selective removal of chromium was responsible for the formation of voids. This supports thermodynamic predictions.* 1854

INTRODUCTION

THE FACTORS WHICH INFLUENCE corrosion in the melts of salts are basically the same as those operating in aqueous electrolytes, as pointed out elsewhere,[1] but their relative importance can differ greatly between these two types of electrolyte. One such factor is diffusion in the metal. At the temperatures associated with aqueous electrolytes this can normally be neglected, but it can play as important a part in fused salt corrosion as in many 'dry' corrosion reactions occurring at similar temperatures.

An extreme example, in which this factor is of prime importance, is the internal oxidation of alloys. Although most of the cases studied so far have involved gas reactions, there is no obvious reason why internal oxidation should not also occur in certain melts. The mechanism of the two cases should be somewhat similar and will not be discussed now. The main purpose of the present note is to draw attention to a type of corrosion during which a component of an alloy is extracted selectively.

The selective removal of the less noble component of an alloy is well known in aqueous corrosion, e.g. it is known as 'dezincification' when it occurs in brass. It is not clear whether, in aqueous systems, the nobler element is oxidized at some stage of the process and is subsequently plated out, or whether it agglomerates by a surface diffusion process. Whichever happens, bulk diffusion is unlikely to play a major part, and consequently when the nobler element is present in sufficiently high concentration, the alloy as a whole tends to behave as if it were noble. This is in part the

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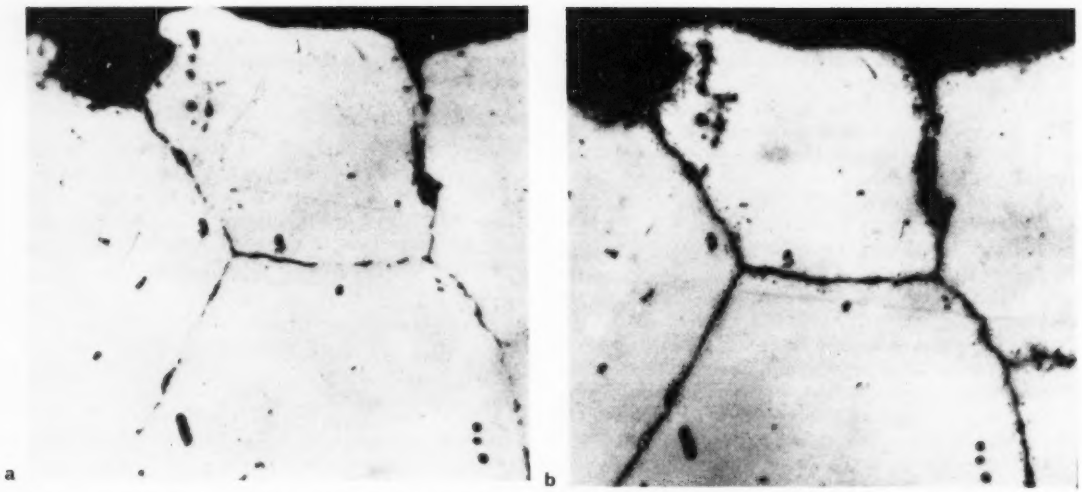


1 *Section through low-carbon 18–8 steel after 116 h in* NaCl–KCl *at 800°C. (a) before etching, (b) after etch in 20%* HNO_3 *for 3 s* ×750

explanation of the so-called 'parting limits' in alloy systems.[2]

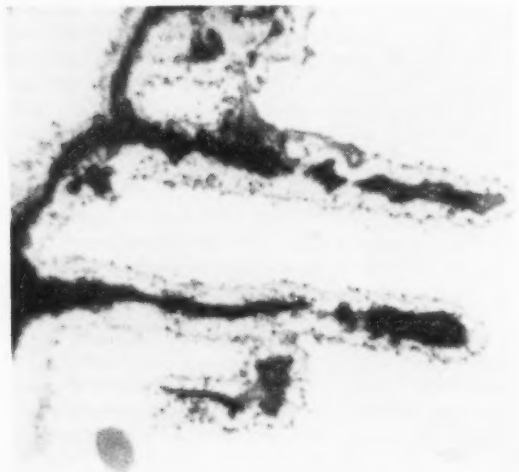
There are a few known cases of aqueous corrosion in which the selective removal does penetrate a great distance into the alloy and the removal of copper from copper–gold alloys by $FeCl_3$ is one of the more spectacular of such cases.[3] Yet there is no reason to believe that bulk diffusion is involved even in this case. However, at the elevated temperatures associated with fused salts, leaching out of a component from the interior of an alloy should be possible and should occur even when the component is only present in a small concentration.

As an example, thermodynamic calculations indicate that in fused chlorides nickel is nobler than iron, while chromium is less so.[4, 1] The data is not entirely reliable, but the following experiment tends to confirm that chromium is the least noble of the metals present in stainless steels.

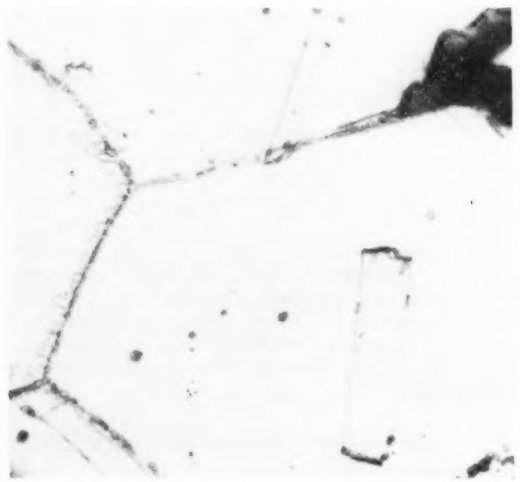
EXPERIMENTAL

If an 18–8 steel is allowed to corrode for about 100 h in a fused 50–50 NaCl–KCl melt at 800°C in the presence of air it is found that it suffers what at first glance may be thought to be ordinary intercrystalline corrosion (Fig.1*a*). More careful examination shows, however, that the attack is by no means continuous at the boundaries and that, especially away from the surface, the specimen merely contains voids which do not appear to be in communication with each other or the outside of the specimen.

On etching the specimen lightly in a solution which should not attack a steel containing 18%Cr (in this case 20%HNO_3) a narrow region on either side of the boundary and next to the outside surface is etched (Fig.1*b*). Examination of these specimens at high magnification reveals that around the larger voids and just below the outer surface there is a second series of very small voids as shown in Fig.2. The material which



2 *High-magnification micrograph showing minor cavities around the major grain boundary voids and slight etching of chromium-poor region* ×2000



3 *Martensite formation at boundaries in low-carbon 18–9 steel after 116 h at 800°C in* NaCl–KCl ×750

is etched by the nitric acid is within the area between the large and small voids.

Etching the material in such a way (e.g. in HNO_3/HCl) that its structure, rather than its composition, is revealed, shows that there is a considerable amount of martensite formation at boundaries (Fig.3). This extends somewhat beyond the region containing easily visible voids.

The simplest interpretation of these results is that chromium is selectively removed by the corrosion reaction and as it diffuses outward, vacancies are moving inwards and segregating to form visible voids (Kirkendall effect). The diffusion rate at boundaries is likely to be very much faster than in the grains so that the chromium in, and adjacent to, the grain boundaries would diffuse relatively rapidly. The formation of voids at boundaries would probably assist the process, since once they are formed, the chromium would diffuse outwards partially by surface diffusion which is a faster process. The voids would grow as fresh vacancies arrive and eventually are likely to join up to give the apparent intercrystalline corrosion.

The martensite formation is consistent with such an explanation since at the 8%Ni level a drop in chromium by a few percent would raise the M_s to above room temperature. The presence of martensite admittedly could be due in part to other factors since an ageing treatment at 800 C does in itself tend to raise the M_s of 18–8 steels. Moreover, a similar effect would also be expected if, as seems probable, the steel is decarburized during the corrosion reaction.

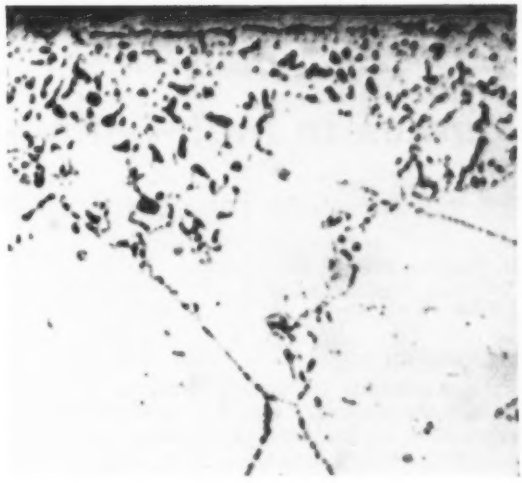
The difference in diffusion rates at boundaries and within grains is normally a function of temperatures, and marked differences in appearance are to be expected depending on temperatures or on alloy systems. In high-nickel alloys the voids within the grains are much more marked (Fig.4). In the case of corrosion by fluorides and caustic, which give a similar effect, it has been shown by analytical means that chromium is in fact removed selectively from the high-nickel alloys.[5, 6]

No conventional analytical work has been done on the specimens shown in Figs.1–3 but it has been possible to show with the aid of the X-ray scanning microanalyser[7] that the material near the boundary was chromium-poor and iron-rich compared with the matrix.

DISCUSSION

The formation of voids, as a result of corrosion in fused salts, is only one of a whole series of other effects we may expect at high temperatures because of diffusion in the alloy. The internal oxidation of a component, as stated before, is equally likely and moreover, if a component in the melt is reduced, there will be the possibility of it forming an alloy. That this can occur is again well known since processes such as 'chromizing' make use of this type of reaction.

Reactions leading to carburization, decarburization, nitriding, etc., are already in this same class and it



4 *Void formation in Inconel after a similar treatment as specimens 1–4* ×500

seems therefore that, in fused-salt corrosion, we have to expect some of the phenomena we normally associate with dry corrosion, as well as some of those associated with the normal electrochemistry of the process.

The void formation described here seems to be the most pernicious type of corrosion to be expected in fused salts. Good creep properties and ordinary oxidation resistance are obvious additional requirements for materials to be used with fused salts but it seems that in developing alloys with suitable overall properties we are somewhat restricted in the selection of alloying metals if we are to avoid this type of failure. Alternatively, in using existing alloys it is necessary to maintain the conditions in the melt sufficiently reducing to avoid the selective removal of the least noble alloying element present.

ACKNOWLEDGMENTS

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Studies in high-temperature gas cleaning

THE REDUCTION OF FUMES FROM OPEN-HEARTH FURNACES

W. Strauss and M. W. Thring

SYNOPSIS

Experiments in filtering waste gases from open-hearth furnaces at temperatures of 375°C, using a coarse granular filter medium, colder than the gases to give thermal precipitation, gave efficiencies better than 90% under many operating conditions and pressure drops varying between 2 and 4 in. water gauge, for a 9-in. thick bed. Beds of 1, 3, and 10½ in. were also investigated under a wide range of operating conditions. Ceramic filters and crushed refractory beds were also tested. 1816

THE PROBLEM

VARIOUS METHODS for dealing with the orange or brown fume produced by oxygen lancing have been investigated; electrostatic precipitators, bag filters, and multicyclones followed by venturi scrubbers have proved successful in individual cases. For the open-hearth furnace in particular, existing processes and possible developments have recently been reviewed by Guthmann[1] and Strauss.[2] Economic considerations which govern the choice of a suitable method for each case are the quantities of dirty gas to be handled, the degree of cleaning required, the rate of gas production, an adequate water supply, and the space available for the erection of gas cleaning plant.

It would be very desirable to have a method of cleaning the waste gases, say with an efficiency of 90% at gas temperature about 600–700°C, in which the unit would be cheaper than those mentioned above, and also have a pressure drop less than about 4 in. water gauge.[3] The pilot plant investigations described here suggest that a bed consisting of coarse granules of refractory material should be able to meet these requirements.

Preliminary studies

At the beginning of the investigation there were no detailed accurate data concerning the physical and chemical properties of the fumes from oxygen-lanced OH furnaces in the duct between the checkers and the waste-heat boiler. This precluded the preparation of representative fume samples in the laboratory, and made it desirable for the tests to be carried out on waste gases drawn from steel furnaces during their normal operation. A preliminary trial indicated that in the steelworks where the tests were to be carried out it was not possible to place the pilot plant on the furnace stage. Here it would have interfered with the normal operation of the furnace, and also services (water, electricity, and compressed air) were not readily available. Below the stage, in the cellar, the services were permanently provided. Temperatures below stage next to the furnace were too high for continuous tests, so the plant was placed 12 ft away at the far side of the cellar, where space was not quite as restricted as close to the furnace.

The apparatus had to be easily assembled and dismantled to facilitate its movement between furnaces, as these, in turn, burned out. The individual sections of the plant had to be small enough to manoeuvre in the constricted space, which also limited the overall size of the pilot plant.

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Sampling fume from the furnace

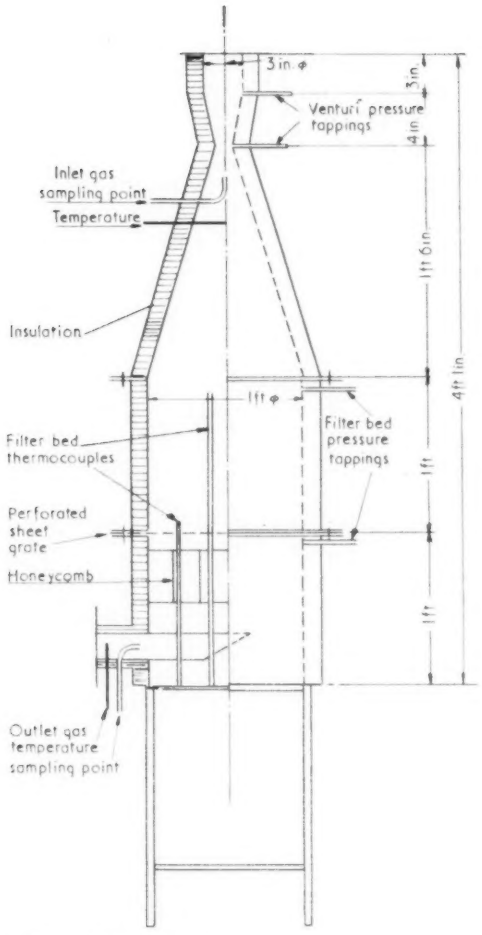
The sampling line from the furnace has been described elsewhere.[4]

Selection of filter media

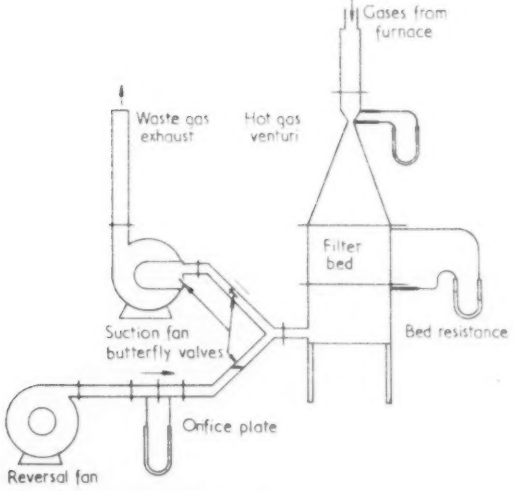
From theoretical considerations the filter material was required to have a rough external surface to present small impaction and interception surfaces to the fume particles passing through the filter. Also the filter media had to stand exposure to the gases at temperatures up to 1000°C to which they could be exposed if used after an inefficient regenerator. High-temperature insulating brick appeared to have these properties, although its microstructure has not been determined. Crushed high-temperature insulating brick graded in the range $\frac{5}{16}$ in.–7 B.S. mesh was used. Finer material was prepared by crushing high-temperature fireclay refractories, a size range of 14–24 B.S. mesh being used in the filter tests. Two types of ceramic filter were also tested. These were Doulton Industrial Porcelains Ltd type CG5 (max. pore size 90–110 μ) and type CG6 (max. pore size 75–90 μ) porcelains, made into discs $2\frac{1}{2}$ in. dia. and $\frac{1}{2}$ in. thick.

Pilot plant for testing the granular refractory

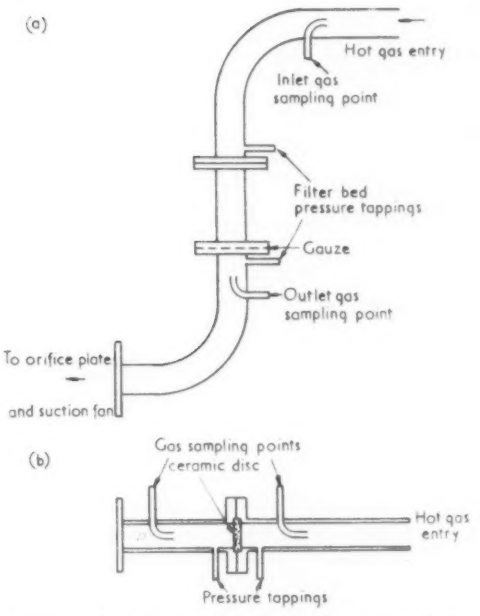
For wall effects to be negligible when considering flow through the filter bed, the bed diameter should be 50 times the granule diameter,[5] and as an average large granule would be about $\frac{1}{4}$ in. dia., a 1-ft dia. bed was chosen. Using a 4000 ft³/h capacity fan, the maximum flow velocity in the free space above the bed would be 1·4 ft/s, giving a value of 5250 R. A dimensioned drawing of the plant is given in Fig.1 and a flow sheet of the assembled plant is given in Fig.2. The plant is described in detail elsewhere.[6]



1 *Pilot plant filter bed unit*



2 *Pilot plant equipment flow sheet*
(a) Crushed refractory retained on gauze
(b) Ceramic discs held in hollowed-out thick flanges



3 *Apparatus for testing crushed refractory and ceramic discs*

Apparatus for testing crushed refractory and ceramic discs

The units for testing the efficiency of crushed refractory and of ceramic discs as gas cleaning filter media were made up of 2 in. nominal pipe sections, flanged together to make orifice plate sections and entrance lines universally usable. The crushed refractory (14–24 B.S. mesh) was retained on a stainless steel wire gauze held between gaskets in flanges (Fig.3*a*). The ceramic discs were held in hollowed-out thick flanges in the 2-in. line, and were sealed at the side with high-temperature cement to prevent short circuiting (Fig.3*b*). Pressure tappings to measure the filter resistance and gas sampling lines before and after the filters were provided. The sampling techniques were the same as used for the pilot plant. The flow rates were measured with a British Standard orifice plate placed after the filter bed. The suction fan used for the pilot plant above was also used here. Although able to handle large volumes against a small resistance, the fan proved unsatisfactory when the resistance exceeded 14 in. water gauge.

Outline of experiments

External factors which influence the dust removal efficiency of a filter bed are the mass flow rate and the velocity of the gases, the dust and fume burden in the gases, and the gas temperature. Internal factors are the nature of the filtering medium, the depth of the filter bed, and the character of the dust deposited in the bed. In the present experiments the filtering medium, its thickness, and the mass flow rate of the gases could be controlled by the operator; the dust burden and the temperature of the gases depended on the phase of steelmaking being carried on in the furnace; while the bed resistance was dependent on the flow rate of the gases and the extent to which the filter bed had been blocked by the deposited dust and fume. The filtration

TABLE I Results of pilot plant experiments—individual test with bed 1–10½ in. of crushed high-temperature insulating brick graded in the range ⅜ in.–7 B.S. mesh

Test no.	Furnace operation	Test time, min.	Mass rate of gas flow, lb/ft²/h	Gas velocity, ft/s	Gas temperature, °C	Fume concentration, g/nm³	Fe_2O_3 in fume, %	Bed thickness, in.	Filter bed efficiency, %	Efficiency in Fe_2O_3 deposition, %	Max. bed resistance, in. water guage
3	R	13	226	1·43	230	1·63	—	10½	90·5	—	2·20
4	R	15	204	1·20	230	4·14	—	10½	96·3	—	2·19
5	C	19	225	1·47	275	0·123	—	10½	87·3	—	2·60
8	O	4·88	172	1·19	290	6·26	—	10½	87·6	—	3·5
9	R	13·13	145	0·96	270	3·83	—	10½	94·7	—	4·0
10	O	6·95	140	0·79	237	6·24	78·0	9	85·0	87·3	4·7
12	R	15·1	214	1·67	345	0·93	64·0	9	90·2	96·0	2·20
14	A	14·4	212	1·54	303	0·54	20·0	9	84·0	97·1	2·40
15	T	12·7	204	1·52	317	0·60	23·0	9	87·5	98·1	2·50
32	M	24·9	126	0·92	305	0·127	36·0	9	84·6	85·7	4·90
36	M	20·4	161	1·26	352	0·163	44·2	9	74·3	74·1	3·60
37	R	18·3	200	1·50	300	1·24	55·2	9	94·2	—	3·90
41	R	20·6	338	3·35	520	1·73	—	3	85·3	—	2·10
45	R	13·0	206	1·29	225	0·595	—	3	65·5	—	0·70
49	F	10·4	317	2·94	460	0·232	—	3	82·7	—	2·35
56	R	15·1	256	2·09	370	0·573	—	1	59·3	—	0·38
71	R	22·2	314	2·51	360	1·18	—	1	87·0	—	0·85
80	R	15·9	354	2·30	455	4·94	—	1	88·2	—	0·52

efficiency was, therefore, measured at several bed thicknesses and as wide a range of gas flow rates as possible. However, when the bed was 9 in. or more deep, the capacity of the suction fan was not large enough to overcome the bed resistance at high flow rates, and the maximum variation in overall gas velocity was from 0·66–1·71 ft/s. Numerous tests were carried out with a 1-in. bed, where the overall gas velocities could be increased to almost 4 ft/s, in an attempt to ascertain the importance of the interception and inertial impaction mechanisms.

Initial tests showed that the dust burden had an important bearing on plant efficiency. Subsequently, attempts were made to run each test at a constant mass flow of gas through the bed. Instantaneous measures of dust burden, which are based on optical obscuration, cannot be readily used with accuracy for measuring plant efficiency, so the filtration method, which gives an average measure of the dust burden over the testing time, was used. Extended processes such as melting and refining were measured in periods based on a single reversal of the regenerators, while short processes such as oxygen lancing were timed over the actual process. As the particles were almost wholly in the sub-micron sizes,[4] elutriation techniques could not be used to determine the 'grade efficiencies' of the filter bed.

Efficiency measurements were made both with the bed initially cold, and with the bed previously heated. Unfortunately, as no external heating for the bed was provided, it was never possible to measure the bed efficiency when the bed was hotter than the gases, to provide conclusive evidence for a thermal precipitation mechanism. Tests were also made with the gases in reverse flow, to see whether the dust and fume particles are blown back into the furnace. The rate of reverse flow was varied to find the rate of heat loss by the bed, and to determine its regenerative capacity.

The 'life' of the filter bed was also estimated. The media were considered exhausted when a single reversal under operating conditions would increase the bed resistance above 4 in. water gauge.[3] Such an estimate is likely to be conservative, as in practice the bed could always be cleared with a powerful localized steam or compressed air jet. Extensive tests were only carried out on the pilot plant, while feasibility studies were extended to the other filter media, efficiencies of these being measured at a few typical operating characteristics.

Results

Table I gives the results of some typical runs using coarse granules, and Table II shows results using finer granules or ceramic filters.

A collection efficiency per unit depth of bed, has been calculated from the equation:

$$\text{total efficiency} = 1-(1-\eta)^n$$

where n = depth of bed in inches. Average values for beds of different thicknesses are listed in Table III.

The statistical significance of the independent variables; mass flow rate, gas temperature, and inlet concentration, was tested in the cases of the 1-in. and 9-in. beds, where the comparatively large number of experimental results (i.e. the greater number of

TABLE II Results of gas filtration tests with crushed firebrick (graded in the range 12–25 B.S. mesh)—1 in. thick

Test no.	Test time, min.	Mass rate of gas flow, lb/ft²/h	Gas velocity, ft/s	Gas temperature, °C	Fume concentration, g/nm³	Bed thickness, in.	Filter bed efficiency, %	Max. bed resistance, in. water guage
94	53·5	262	1·10	60	0·107	1	45·0	14·0
95	18·8	216	0·92	63	—	1	—	12·5
96	24·0	216	0·92	63	—	1	—	12·5
Results of tests on ceramic filters—Doulton Industrial Porcelain Ltd Types CG6 and CG5								
		400	1·68					
97	8·27	160	0·67	60	0·196	CG6	75·6	13·6
98	3·0	215	0·90	60	0·086	CG5	57·3	13·6
99	12·16	230	0·97	60	0·035	CG5	57·1	13·6

Fe_2O_3 was not determined in the fume

TABLE III Average values for efficiency of fume collection, mass flow of gas, gas velocity, and temperature from experimental data for pilot plant with various depths in the range $\frac{1}{16}$ in.–7 B.S. mesh crushed high-temperature insulating brick

	Bed thickness, in.			
	1	3	9	10·5
Overall pilot plant efficiencies, %	68·1	74·4	86·4	89·7
Efficiencies for filter bed only, %	56·4	67·7	78·7	88·4
Specific efficiency per inch depth, based on average efficiencies	56·4	31·4	15·6	18·5
Specific efficiency per inch depth, based on similar conditions	44	—	25	—
Average mass flow, lb/h/ft²	273	279	174	201
Average temperature, °C	360	355	268	251
Average velocity, ft/s	2·18	2·22	1·19	1·33
Average inlet concentration, g/nm³	1·54	1·47	1·36	2·37

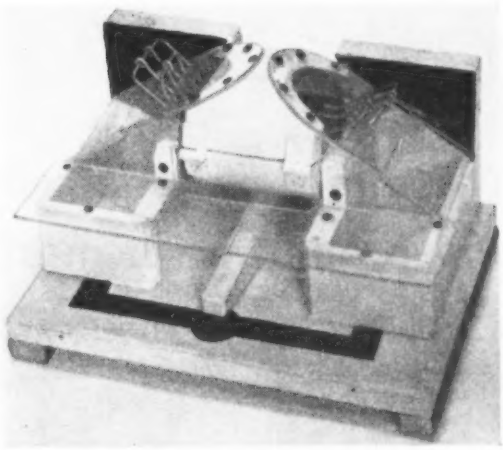
degrees of freedom) made this possible. In both cases the inlet concentration term proved highly significant. In the 9-in. bed the mass flow rate and the temperature proved significant, but only at the 5% level. In the 1-in. bed the mass flow rate was significant at the 1% level, and the temperature was just within the 5% level.

The 'life' of the bed was found to be expressed roughly by a dimensionless factor β in a modified form of the Ergun and Orning equation.[7] For a new bed, β was about 8·5 at 300°C, while for a nearly exhausted bed β was about 20.

The mechanisms involved in the filtration are discussed in more detail elsewhere.[6]

The applications of this technique to OH steel furnace fume will now be considered.

Figure 4 shows a model of a fuel fired, oxygen lanced, hot metal OH furnace specially designed to make use of this type of combined high efficiency heat exchange and fume removal. This model is designed entirely on aerodynamic lines; for example, the gases leave the furnace through a long narrow slot which gives the best conditions for turning the corner, and also enables a row of oil burners and oxygen lances to be arranged across each end of the square bath without giving a very great length of water-cooled surface exposed inside the furnace. There are no slag pockets and the gases go straight round this corner into the large cross-sectional area bed which is about 3 ft thick, consisting of three layers, the top one of half bricks,



4 *Model of fuel fired oxygen lanced hot metal OH furnace*

the middle one of 1-in. particles, and the bottom one of the fine particles about 1 ft thick used in these experiments. The whole is supported on a cast-iron grate and is arranged so that it can be removed on a trolley and replaced by a clean bed once a week without appreciable interruption of operation. The bed is filled with cheap material by mechanical means and thus is not expensive to replace. It is too early yet to assess the economics of this system, but it seems likely that it would be considerably cheaper than the combination of a classical regenerator with waste-heat boiler and electrostatic precipitator, while the pressure drop will be appreciably less than with the venturi scrubber, and thus the power to operate will be less.

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The effect of composition on the structure and properties of martensite

K. J. Irvine, B.Sc., Ph.D., F. B. Pickering, A.Met., A.I.M., and J. Garstone, M.Met., Ph.D.

SYNOPSIS

An electron microscope study has been made of the structures of plain carbon steels containing carbon contents up to 0·9%C and of alloy steels with a similar range of carbon contents but with single additions of up to 4% of Cr, Mn, Ni, *and* Mo. *These steels have been examined in the as-quenched condition and also after tempering at temperatures between 100° and 300°C. The mechanical properties of these steels have been determined so that the main structural features can be related to the mechanical properties obtained.*

From this work it has been concluded that the major factor controlling the hardness and strength of martensite is the amount of carbon in solid solution. The fine grain size of the structure does produce an increase in strength compared with a polygonal ferrite structure, but there is no variation in grain size with increasing carbon content, so the effect of grain size is relatively small in a high-carbon steel. If grain size can be reduced, for example, by warm working before martensite transformation, strength can be appreciably increased.

Alloy elements have little effect on the martensite structure, but mainly affect the tempering process, retarding carbide precipitation and maintaining the high martensitic hardness to higher temperatures. In low-carbon steels an alloying element may reduce autotempering and hence significantly increase hardness, but in general the effect of an alloying element is relatively small.

The mechanical properties of plain carbon martensite are most satisfactory with between 0·3 and 0·5%C and when tempered at 200–300°C. Alloy steels at any particular strength level have better ductility and also very much better impact properties, particularly when molybdenum is present. This appears to be due to a modification of carbide precipitation so that films at martensite plate boundaries are eliminated. 1881

EFFECT OF COMPOSITION ON STRUCTURE AND PROPERTIES OF MARTENSITE

THERE IS CONSIDERABLE INTEREST in the properties of martensite and in the way these are affected by composition and structure, because the most common method of achieving a high level of tensile strength in a steel is quenching to produce martensite and subsequently, tempering to achieve adequate ductility. It is true that the martensitic structure is rarely used untempered because with low-temperature tempering the combination of strength, ductility, and toughness is very much improved. This situation also exists, however, with an alloy which is hardened by a precipitation reaction involving either a carbide or intermetallic compound. Such an alloy is rarely used in the peak hardened condition, but preferably with some degree of overageing so that the properties, other than hardness, are improved.

Therefore, in this paper the effect of composition and structure has been studied on the properties of martensite both as-quenched and also lightly tempered (i.e. up to 300°C). No attempt has been made to deal with the structures produced by tempering the martensite at higher temperatures because most of the martensitic hardness is then lost and the properties depend upon the size and distribution of carbide particles formed during tempering.

Generally speaking, alloying elements are introduced for purposes other than modifying the martensitic structure. They may be added to improve hardenability, tempering resistance or corrosion resistance, with the underlying assumption that the martensite produced is relatively unaffected. However, one primary object of this paper is to show the effect of composition on the martensitic structure itself.

Although a considerable amount of information has been published about the martensite transformation, there is little information on the microscopic structure and mechanical properties. Recent papers by Busby *et al.*[1] and Aborn[2] have reviewed the properties of low-carbon martensite. These have shown some quite attractive properties but have brought out the fact that because of the high M_s temperature of these steels there is autotempering and the structures show precipitates of Fe_3C or ϵ-carbide even with the fastest rates of cooling.

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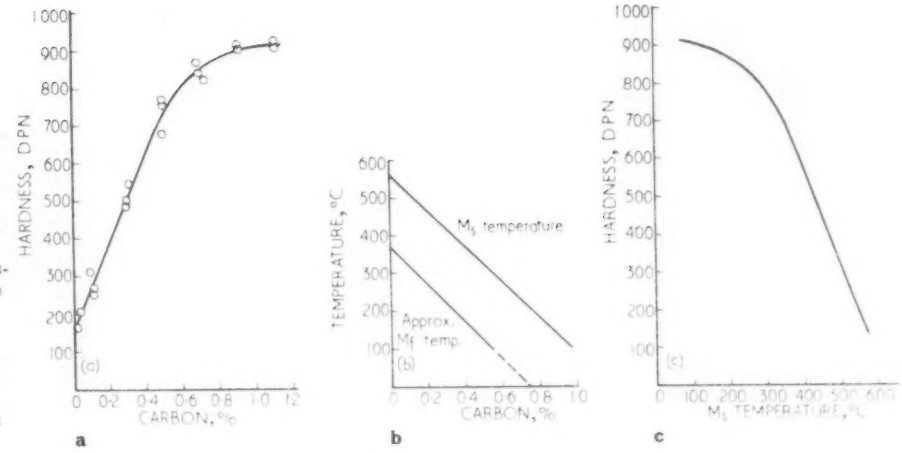


1a *Effect of C on the hardness of martensite in plain carbon steels*

1b *Effect of C on the martensite transformation temperatures (after Steven and Haynes)*

1c *Relationship between M_s temperature and hardness in plain carbon martensite*

There are in fact, many difficulties in investigating the properties of martensitic structures, particularly in plain carbon steels. Firstly, the hardenability is low so that at low carbon contents it is difficult to obtain a fully martensitic structure free from bainite. Also at higher carbon contents there is likely to be some difficulty with quench cracking and with retained austenite in the highest carbon content steels. These features and the inherent low ductility of higher carbon martensitic structures makes the interpretation of tensile properties difficult. However, a lot of information can be gained from hardness determinations and improved properties can be obtained as soon as some tempering is carried out.

When alloying elements are used there is some difficulty in differentiating between several effects. For instance, an alloying element may produce solid-solution hardening, and since the hardness of the steel can be measured quite easily it might be thought that the solid-solution hardening effect could therefore be assessed. It must be remembered however, that the alloy element also depresses the M_s temperature and in a low-carbon steel this can result in less autotempering and hence a higher hardness because the carbon content of the martensite is effectively higher. The alloy element also increases the hardenability and in low-alloy steels part of the initial increase in hardness may be due to a more efficient quench to martensite. Some care must be taken in interpreting the results, therefore, particularly when small additions of alloying elements are being made to low-carbon steels.

Plain carbon martensite

Since martensite is a supersaturated solution of carbon in body-centred α-iron it is to be expected that carbon will have a major effect on the structure and properties. The most important effects are:

Effect of carbon on hardness

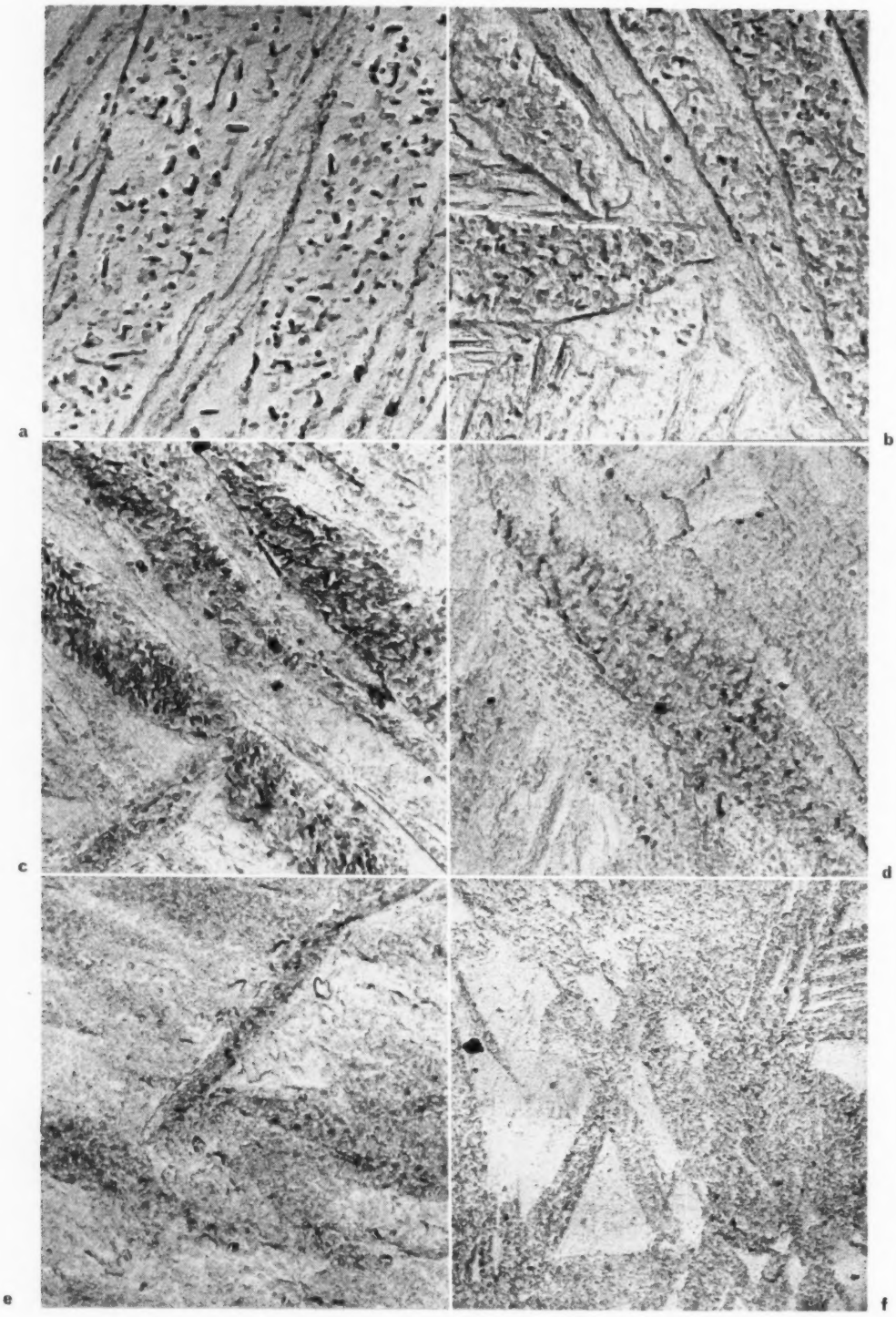
Carbon is the element which has the major effect in increasing the hardness of martensite as shown in Fig.1*a*. With increasing carbon the hardness increases quite sharply in a linear manner, but above about 0·7%C, the rate of increase of hardness falls off considerably.

Effect of carbon on the M_s

It is possible to calculate the M_s temperature according to the formula of Steven and Haynes,[3] and when this is done for the series of plain carbon steels the relationship between carbon and M_s shown in Fig.1*b*, is obtained. Steven and Haynes showed that the M_f is about 215°C below the M_s and the martensite transformation range is also indicated in Fig.1*b*. It is clear that above 0·7%C the martensite transformation may be incomplete and therefore the maximum hardness will not be obtained. When the quenched hardness of the steels is expressed against the M_s temperature as in Fig.1c it is clear that the uniform increase in hardness with increasing carbon is only interrupted when the M_s is depressed to about 200°C at which stage incomplete martensite transformation will be obtained.

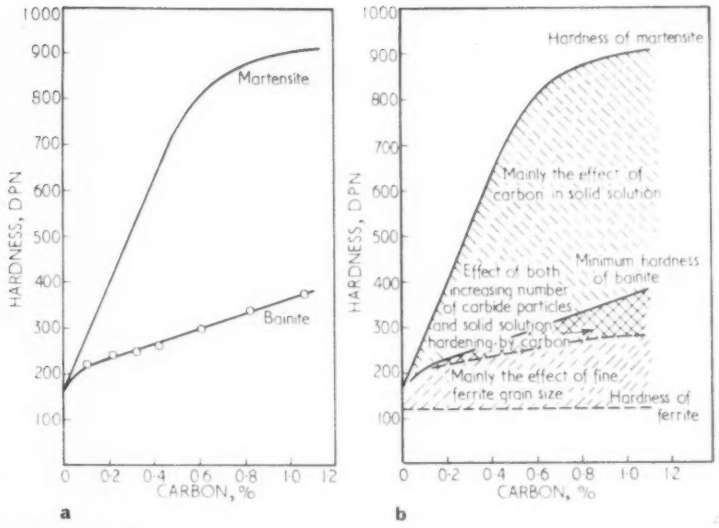
Effect of carbon on microstructure

The structures of the quenched plain carbon steels were examined using both optical and electron microscopes. The optical microstructures showed that the martensite grain size was small, but there was little change with increasing carbon. At the lower carbon contents (up to 0·4%), the martensite etched quite darkly which was probably due to precipitation within the martensite plates. Above 0·4%C the etching became lighter so that the individual martensite plates could not be seen clearly. The electron microstructures are shown in Fig.2. The 0·1%C steel was partially bainitic because of the low hardenability and therefore it is a selected area showing a martensitic structure which is illustrated in Fig.2*a*. There has been extensive autotempering because of the high M_s temperature and consequently the martensitic plates contain many precipitated carbide particles. The structure is more uniform in the 0·2%C steel but there has been extensive autotempering of martensite plates (Fig.2*b*). The martensitic structure is not so clearly defined in the 0·3 and 0·5%C steels (Figs.2*c* and *d*) and there are few well-developed carbide particles. However, there is extensive roughening of the martensite plates which might indicate preprecipitation zones. The 0·7 and 0·9%C steels are very light etching and the main structural feature is a roughening of the martensite on etching which helps to define the martensite plates (Figs.2*e* and *f*). The 0·9%C steel shows some white areas which are assumed to be retained austenite.



a 0·1%C *b* 0·2%C *c* 0·3%C *d* 0·5%C *e* 0·7%C *f* 0·9%C

2 *As-quenched structures of plain carbon steels* ×15000


HARDNESS, DPN
CARBON, %
Martensite
Bainite
a
Hardness of martensite
Mainly the effect of carbon in solid solution
Effect of both increasing number of carbide particles and solid solution hardening by carbon
Minimum hardness of bainite
Mainly the effect of fine ferrite grain size
Hardness of ferrite
b

3a *Effect of C content on minimum hardness of bainite compared with martensite*

3b *Factors affecting strength of martensite and bainite*

Modern theories of martensite formation have shown that the martensite transformation is the result of a shear-type transformation caused by the movement of dislocation walls through the austenite structure. This often leaves a high residual stress concentration and also a high residual concentration of dislocations in martensite. There will also be an intense strain field surrounding a carbon atom in an interstitial site in the martensite lattice. The anisotropy of these strain fields distorts the body-centred cubic lattice and the amount of distortion increases with increasing carbon. The effect of these strain fields associated with the carbon atoms can be described as solid-solution hardening due to carbon. It is to be expected therefore, that these internal strains will have a major effect on martensite hardness.

Since there is little evidence of much change in grain size with increasing carbon it seems likely that the fine grain size of martensite introduces a hardness increment compared with polygonal ferrite, but that this is not responsible in any way for the major increase in hardness which occurs when the carbon content is raised.

It is difficult to determine the effect of precipitation in a martensitic structure. Certainly in low-carbon steels, autotempering occurs because the M_s temperature is high. This could decrease the hardness because carbon is removed from solution, but the effect might be compensated by a precipitation hardening effect. With increasing carbon content there is a lower M_s temperature and hence less autotempering, so that the carbon content is more fully effective. On the other hand, there are fewer precipitated carbides and the effect of the preprecipitation zones is difficult to assess. On balance however, it does not seem that the presence of precipitates is a major factor in the hardness of martensite.

Comparison of bainitic and martensitic structures

Since the important factors influencing the hardness of a martensitic structure are grain size, carbide particle dispersion hardening, and solid-solution hardening by carbon, it was interesting to compare the martensitic structure with the bainitic structure where the same three factors control the hardness. To make a direct comparison between these two structures it was necessary to use a steel composition which would give bainitic structures at air-cooling rates. It was known that a $\frac{1}{2}$%Mo–B composition was suitable and with this base a series of steels with varying carbon was made and tested in both the air-cooled and water-quenched conditions. This slight alloy addition does not affect the martensitic structures appreciably but does enable fully bainitic structures to be obtained. The effect of carbon on the hardness of these two structures is shown in Figs.3*a* and *b* from which it is possible to assess the importance of the different factors influencing hardness.

With regard to low-carbon bainites, a detailed description has already been published[4] which shows the effect of grain size and carbide particle density, and which indicates that the hardness of a low-carbon low-alloy bainite is mainly due to the fine grain size. Work has also been completed on medium-carbon bainites which will be published separately, but a brief summary of this is that increasing carbon depresses the transformation temperature, which has the effect of decreasing grain size and increasing the number of precipitated carbide particles. The carbide particle density is also increased simply due to the higher carbon content itself. It is concluded therefore, that in the low carbon content steels (0·1–0·2%), the higher hardness of the bainitic structure compared with polygonal ferrite is mainly due to the considerably finer grain size. This effect is slightly increased in the higher carbon steels, but there is also an additional hardness increment due to the increased number of carbide particles.

It has been shown that the martensitic grain size is not very different from the bainitic grain size (in fact, the martensite is slightly coarser) and is unaffected by increasing carbon. Also except at low carbon contents there are few carbide particles precipitated. Consequently, it is suggested that the major reason for the difference in hardness between the bainitic and martensitic structures is the solid-solution hardening effect of carbon. These effects are summarized in Fig.3*b* which indicates the relative importance of the

TABLE I Mechanical properties of plain carbon steels

Tempering treatment	Quenched	100°C 1	100°C 2	100°C 7	200°C 1	200°C 2	200°C 7	300°C 1	300°C 2	300°C 7
0·2%C steel										
Tensile strength, tons/in²	95·2	104·0	99·8	109·6	94·8	112·0	94·0	88·0	87·0	87·0
0·2% proof stress, tons/in²	82·4	86·4	86·4	94·4	80·0	102·4	80·0	75·2	80·0	72·0
Elongation, %	5	7	4	6	9	8	6	8	9	9
Reduction of area, %	17	16	10	9	27·0	16	11	25	31	30
Impact, ft lb	...				...					
DPN	446	487	326	444	457	449	446	315	325	357
0·3%C steel										
Tensile strength, tons/in²	102·4	102·8	97·6	104·0	92·8	92·8	94·4	82·4	83·2	80·4
0·2% proof stress, tons/in²	88·0	87·2	82·4	88·8	72·8	78·4	82·4	68·0	77·6	73·6
Elongation, %	4·5	5·5	5·5	7·0	7·5	6·0	7·0	9·0	10·5	10·0
Reduction of area, %	7·0	7·0	13·0	15·0	16·0	15·0	22·0	28·0	35·0	34·0
Impact, ft lb	8	8	9	8	9	10	11	10	11	12
DPN	564	554	602	517	530	511	502	444	432	420
0·5%C steel										
Tensile strength, tons/in²	36·4PF*	60·0PF	56PF	86PF	80PF	125·6	132	107·2	103·2	103·6
0·2% proof stress, tons/in²	...	...	...	...	...	114·4	108·8	96·8	90·4	91·2
Elongation, %	...	...	...	...	...	1·0	4·0	7·0	8·0	7·0
Reduction of area, %	...	...	...	...	...	2·0	4·0	22·0	21·0	23·0
Impact, ft lb	4	4	5	4	6	6	5	5	6	6
DPN	680	695	730	666	657	594	571	499	467	470
0·7%C steel										
Tensile strength, tons/in²	16·8PF	...	...	24PF	31·6PF	...	...	56PF	66·4PF	64·8PF
0·2% proof stress, tons/in²	...	...	...	...	...	...	...	...	...	...
Elongation, %	...	...	...	...	...	...	...	...	...	...
Reduction of area, %	...	...	...	...	...	...	...	...	...	...
Impact, ft lb	2	...	...	2	3	3	3	4	3	4
DPN	823	867	862	849	786	736	661	564	516	547

* PF = Premature fracture

different features and emphasizes the major effect of carbon in solid solution.

There is a tendency for these two structures to be even more similar at low and high carbon contents. At low carbon contents in the quenched condition auto-tempering will produce many precipitated carbide particles so that the martensitic hardness is reduced almost to that of bainite. At high carbon contents, in the transformation to bainite, there is an increasing tendency for carbon to be held in solid solution with the result that the hardness is greater than would be expected and approaches that of the martensitic structure.

Mechanical properties of plain carbon martensites

Up to 0·2%C the martensitic structure is difficult to achieve in specimens large enough to determine mechanical properties due to lack of hardenability. Above 0·5%C the specimens are so brittle that premature failure always occurs unless the tempering temperature is above 300°C. Consequently, the mechanical properties of plain carbon martensites are largely confined to carbon contents between 0·2 and 0·5%. These properties are given in Table I. The steels used had a base composition of 0·5%Mn, 0·2%Si.

In the quenched condition the tensile strength seems to be governed by the available ductility and only the 0·2%C steel has a strength level which compares reasonably with the hardness. Tempering at 100°C does not cause any significant reduction in the strength of the 0·2 and 0·3%C steels and the ductility is slightly increased. An increase in strength is obtained from the higher carbon steels, although there are still premature failures. Tempering at 200°C does not reduce the strength level of the 0·2%C steel, although the ductility is increased. The strength of the 0·3%C steel is beginning to drop with tempering at 200°C and it will be seen that there is a better combination of strength and ductility when the tempering time is increased to 7 h. The properties of the 0·5%C steel are also improved by tempering at 200°C, although the ductility figures are rather low. Premature failure is obtained in the 0·7%C steel. Tempering at 300°C produces appreciable softening in the 0·2, 0·3, and 0·5%C steels, and the ductility values are reasonable, although still slightly low for these tensile strength levels. Premature failure is still obtained in the 0·7%C steel.

Charpy impact tests were carried out on each of these steels in the quenched and tempered conditions. There was surprisingly little variation with tempering temperature but carbon content had a noticeable effect. This effect can be summarized as follows:

0·3%C 9–12 ft lb
0·5%C 4–6 ft lb
0·7%C 2–4 ft lb

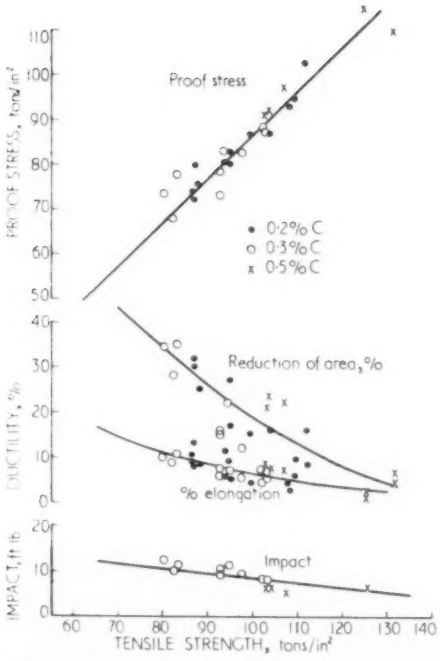
It is possible to consider these mechanical properties together and show how the other mechanical properties are related to the tensile strength. This relationship is shown in Fig.4 and it is clear that the variation in carbon does not affect the general relationship that increasing tensile strength is accompanied by increasing 0·2%PS and decreasing ductility and impact.

Hardness changes during tempering

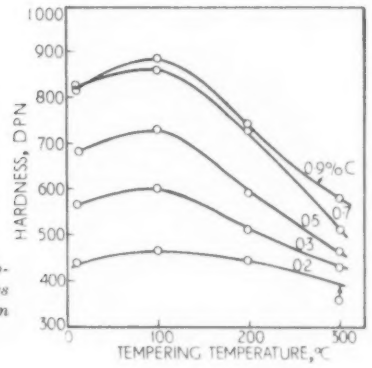
The change in hardness of the different plain carbon steels after tempering the martensitic structure is shown in Fig.5. There is a marked increase in hardness after tempering at 100°C, which increases in magnitude with increasing carbon content, and there is then rapid softening as the tempering temperature increases. The higher-carbon steels soften more than the lower-carbon steels, so that the hardness difference of about 300 DPN between the 0·3 and 0·9%C steels after tempering at 100°C, is reduced to 150 DPN after tempering at 300°C.

Microstructural changes during tempering

0·1%C Although this steel was not completely hardened, it was possible to examine the martensitic



4 *Mechanical properties of lightly tempered plain carbon martensites*



5 *Effect of low-temperature tempering on hardness of varying carbon martensite (for 2-h tempering)*

areas. The quenched structure contained autotempered carbides, and the number of these carbide particles increased slightly with tempering up to 300°C. One important feature was that there was precipitation of carbide films along the martensite plate boundaries.

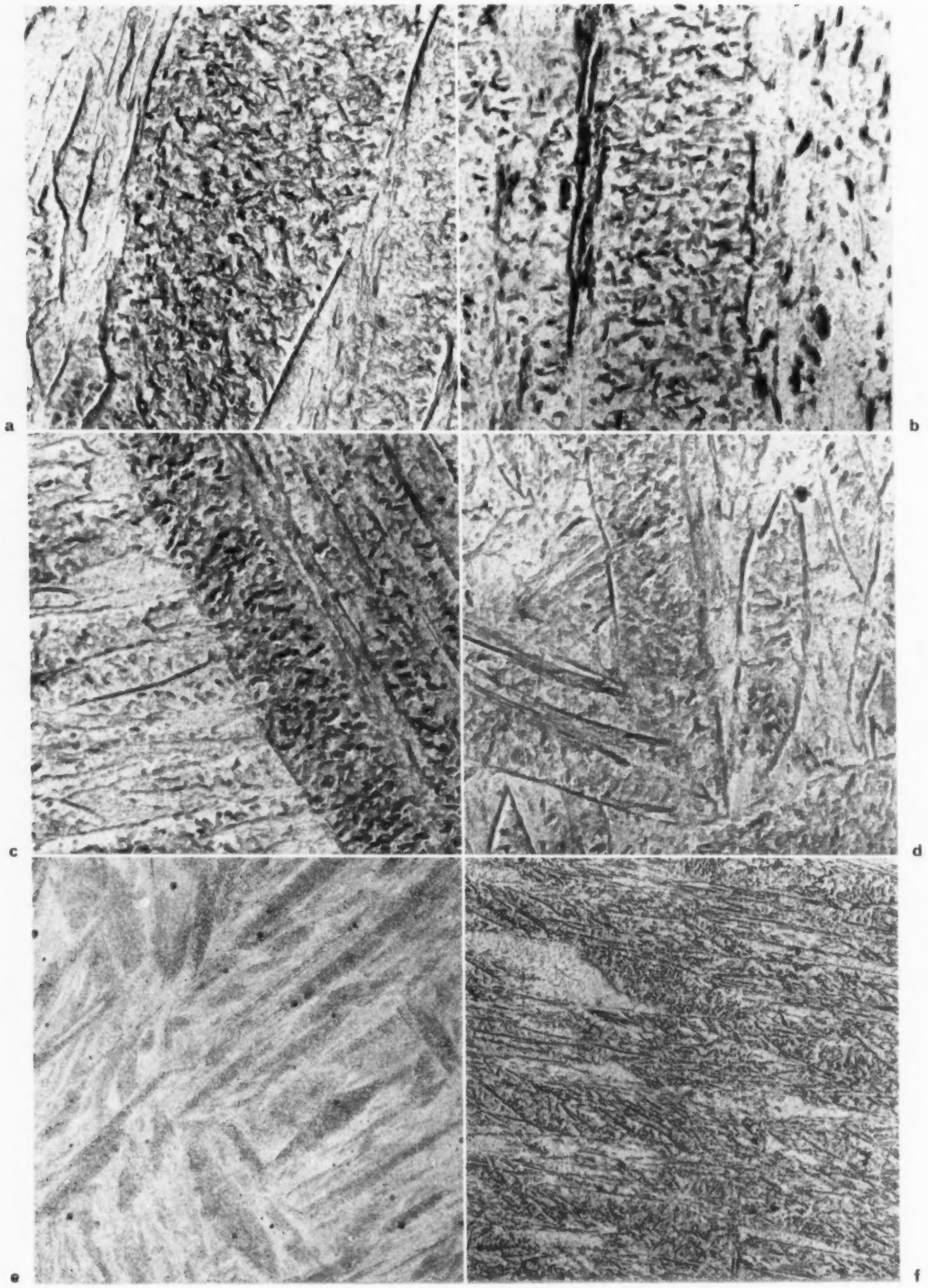
0·2%C The as-quenched structure contained many carbide particles within the martensite plates (Fig.2*b*). This precipitation was intensified by tempering. Precipitation of some very small carbide particles occurred at 100°C and was quite marked after 7 h (Fig.6*a*). The precipitate coarsened at 200° and 300°C and there were films of carbide formed around the martensite plate boundaries (Fig.6*b*).

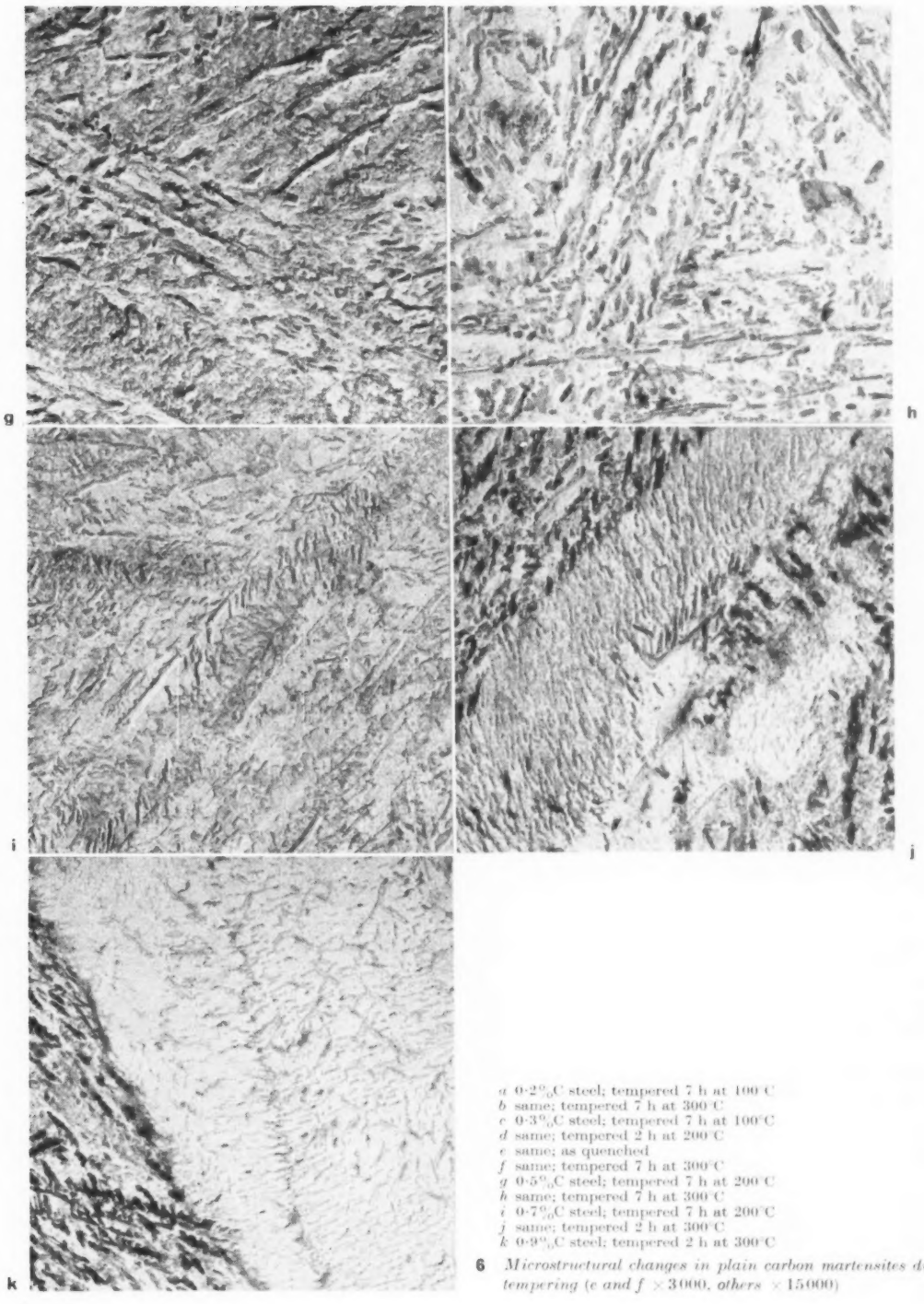
0·3%C The as-quenched structure contained few autotempered carbides, and these were only in the first-formed martensite plates (Fig.2*c*). There was little change after 1 and 2 h at 100°C but quite definite precipitation of carbide within the martensite needles after 7 h at 100°C (Fig.6*c*). The removal of carbon from solid solution in the martensite by this carbide precipitation explains the drop in hardness and increased ductility observed in the tensile test after this tempering treatment. Similar precipitation within the martensite plates was observed after tempering at 200°C, but there were also carbide films precipitated at martensite plate boundaries, which were quite pronounced after 2-h tempering (Fig.6*d*). Because of this feature in the microstructure it is not difficult to understand why the ductility values are little improved in spite of a marked decrease in hardness. The carbide precipitation became coarser after tempering at 300°C, and the grain-boundary films were not continuous. This fact, together with the decrease in strength resulting from further carbide precipitation, explains why the ductility values have improved. This process of carbide precipitation within the martensite plates during tempering at low temperatures is well illustrated by the low-magnification electron micrographs in Figs.6*e* and *f*. These show the relatively few carbides present in the quenched structure compared with the extensive precipitation after tempering at 300°C. This tempered structure, which is so clearly shown in Fig.6*f* would merely show as rapid etching and darkening under the optical microscope.

0·5%C There were few carbides in the quenched structure and there is little change after tempering at 100°C. It is not surprising therefore, that the tensile properties show no improvement. The lack of carbides in the quenched structure means that there are few nuclei for precipitation, and the high degree of supersaturation existing with this 0·5%C steel means that there is a large number of nuclei forming at the tempering temperature, which consequently will not grow so rapidly. This is well shown by the sample tempered at 200°C for 7 h (Fig.6*g*). By this stage of tempering the hardness has decreased from 680 to 571 DPN, and a small amount of ductility enables a reasonable tensile test to be obtained. Coarsening of the carbide particles occurred during tempering at 300°C, but there was little evidence of carbide films at martensite plate boundaries (Fig.6*h*).

0·7 and 0·9%C steels Very similar structures were produced in these two steels. The high degree of supersaturation produces a large number of nuclei for carbide precipitation and restricted growth at low tempering temperatures. In fact, tempering at 100°C does not produce any marked precipitation of carbide plates and it may be that the general roughening observed is due to ϵ-carbide. There is very heavy precipitation of carbide at 200°C which occurs as plates across the martensite plates (Fig.6*i*), but these carbides were not very well developed. The hardnesses were still so high (>650 DPN) that no tensile properties could be obtained. The main difference after tempering at 300°C was that the carbide plates became better developed (Fig.6*j*) but in spite of the fact that the hardness had now dropped to 550 DPN it was not possible to get satisfactory tensile properties. It was noticed that there was little evidence of carbide films at martensite plate boundaries as had been observed in the lower-carbon steels. The poor mechanical properties may therefore be due to microcracks formed during quenching.

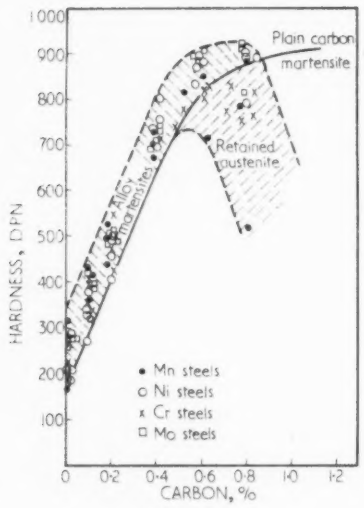
It was in the 0·9%C and higher-carbon steels that some retained austenite was observed in the micro-


a
b
c
d
e
f



a 0·2%C steel; tempered 7 h at 100°C
b same; tempered 7 h at 300°C
c 0·3%C steel; tempered 7 h at 100°C
d same; tempered 2 h at 200°C
e same; as quenched
f same; tempered 7 h at 300°C
g 0·5%C steel; tempered 7 h at 200°C
h same; tempered 7 h at 300°C
i 0·7%C steel; tempered 7 h at 200°C
j same; tempered 2 h at 300°C
k 0·9%C steel; tempered 2 h at 300°C

6 *Microstructural changes in plain carbon martensites during tempering (e and f* × 3000, *others* × 15000)



7 *Effect of alloying elements on hardness of martensite*

structure. This remained unchanged with low-temperature tempering until 300°C was reached. These areas then showed evidence of transformation which was presumably to a low-temperature bainite (Fig.6*k*).

The tempering curves in Fig.5 show an increase in the hardness of each of the steels after tempering at 100°C. High magnification examination of the microstructures of these specimens showed an extremely fine matrix precipitate which may well be ε-carbide and it seems clear that it is this precipitation which is producing the hardness increase.

Effect of alloying elements on martensite

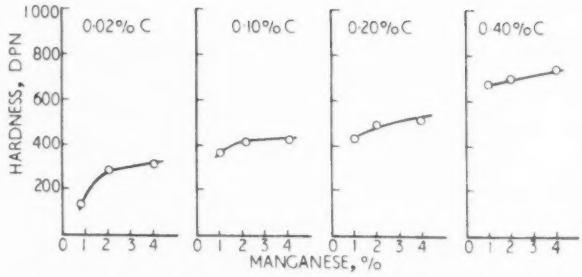
Effect on hardness

To examine the effect of alloying elements on martensite, several series of steels were made with up to 0·8%C content and with single additions of up to 4% of Mn, Ni, Cr, and Mo. (The remaining base composition was 0·5%Mn, 0·2%Si.) These steels were all quenched in iced brine from a standard austenitizing temperature of 1000°C and the hardness values measured. These results are shown in Fig.7 and there are two main features to note. With increasing carbon content, the hardnesses for the alloyed martensites lie in a relatively narrow band above the line for the plain carbon martensites. It would appear therefore, that the effect of alloying elements is relatively small compared with the effect of carbon. The second feature of interest is the tendency to reduced hardnesses at high carbon contents because of increased amounts of retained austenite. This effect is particularly marked with manganese which has a strong effect in depressing the M_s temperature.

An alloying element will have several effects on the martensitic structure and it is suggested that these are:

(i) solid-solution hardening
(ii) depression of M_s and hence reduction in autotempering
(iii) retardation of the tempering process.

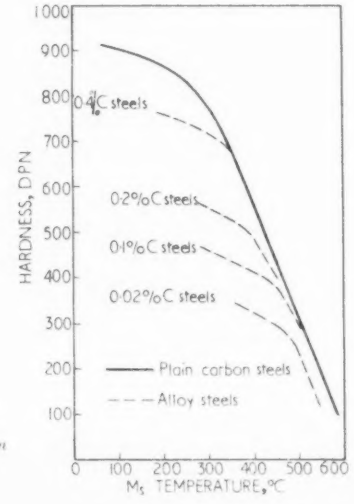
It is possible to see their relative contribution to the hardness increase by examining the effect of an alloying element in more detail. The effect of increasing



8 *Effect of* Mn *on hardness of martensite of different* C *contents*

manganese on the martensitic hardness of the different carbon content steels is shown in Fig.8. In the lower-carbon steels, the first manganese additions up to 2% have more effect in increasing hardness than the subsequent increase up to 4%. This effect is diminished as the carbon increases until with 0·4%C there is a consistent but slight increase in hardness with increasing manganese. It is suggested that this variation in hardness increase is due to the effect of the manganese on autotempering. There is quite an appreciable amount of autotempering in a 0·02–0·10% plain carbon steel so that the true martenistic hardness is not realized. The first addition of manganese reduces this autotempering and since this increases the amount of carbon in solid solution, the effect on hardness is quite pronounced. Once autotempering has been suppressed however, increasing manganese only has a solid-solution hardening effect, which is relatively slight. With a 0·4%C steel, there is no autotempering in the plain carbon steel, so that even the first addition of manganese only has a solid-solution hardening effect which results in the slight but consistent hardness increase with increasing manganese.

While this effect has been discussed for manganese a similar effect can be observed with other alloying elements according to their effect in depressing M_s. The general effect is shown in Fig.9 which illustrates the importance of the M_s in this relationship. The lines representing alloy steels summarize the data for all the steels containing single additions of Cr, Mn, Ni, and



9 *Effect of alloying elements on the relationship between M_s temperature and hardness*

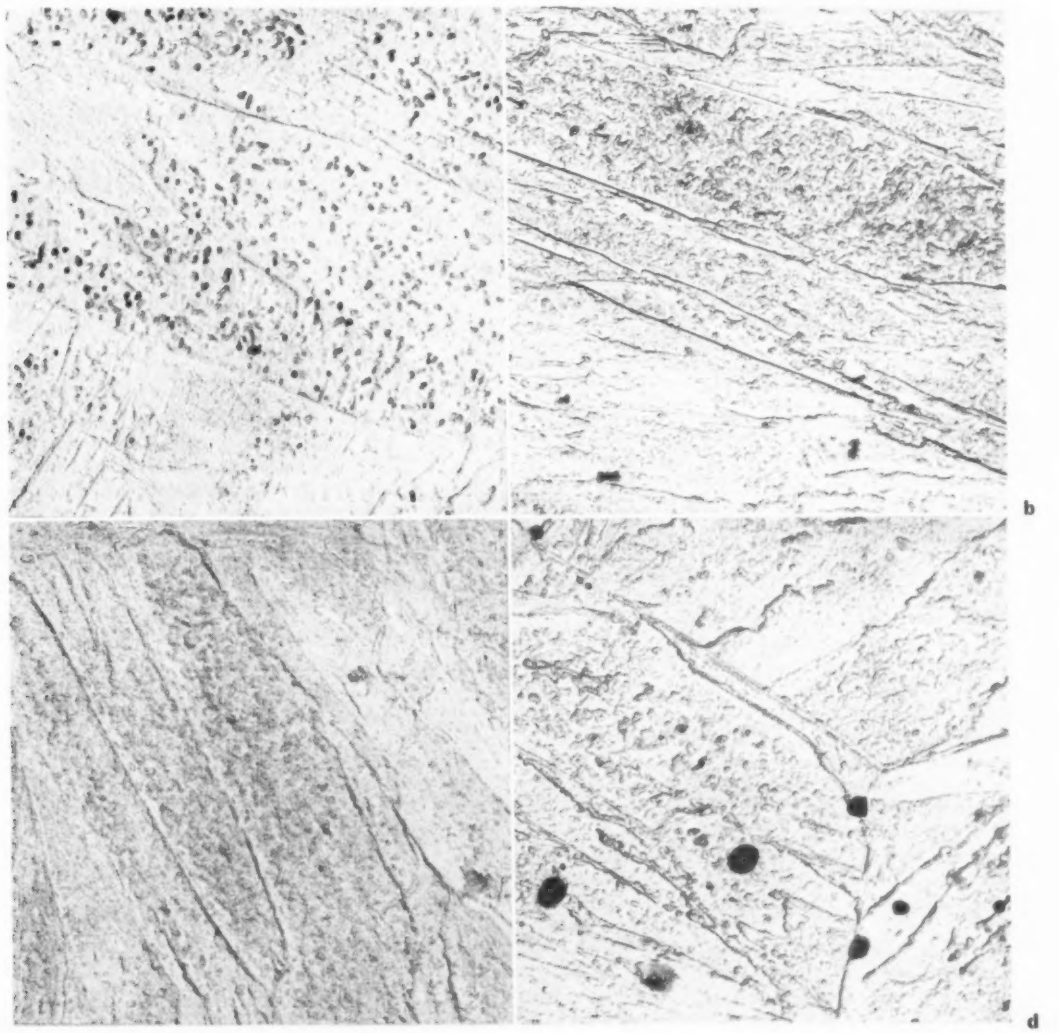
Mo. Compared with the line for plain carbon steels, the alloyed steels depart from this line in a characteristic way. With the first addition of the alloying element, the alloyed steel follows the plain carbon line, which implies that this addition of the alloying element by reducing the M_s has reduced autotempering and in effect raised the carbon content of the martensite. The subsequent addition of the alloying element merely results in solid-solution hardening so that the hardness increase is much less than that produced by increasing carbon. A similar effect is produced by each of the alloying elements, and the initial hardness increase decreases in magnitude with increasing carbon until it is not observed in the 0·4%C steels.

Effect on microstructure

One of the most important effects of alloying elements is that the M_s temperature is depressed and therefore autotempering is suppressed in low-carbon steels. This effect can be clearly seen when the microstructures are examined with the electron microscope. Typical quenched structures are shown in Fig.10 illustrating the effect of Cr, Ni, Mn, and Mo. When these are compared with the quenched plain carbon steel which is shown in Fig.2*b* the marked reduction in the amount of autotempering is seen.

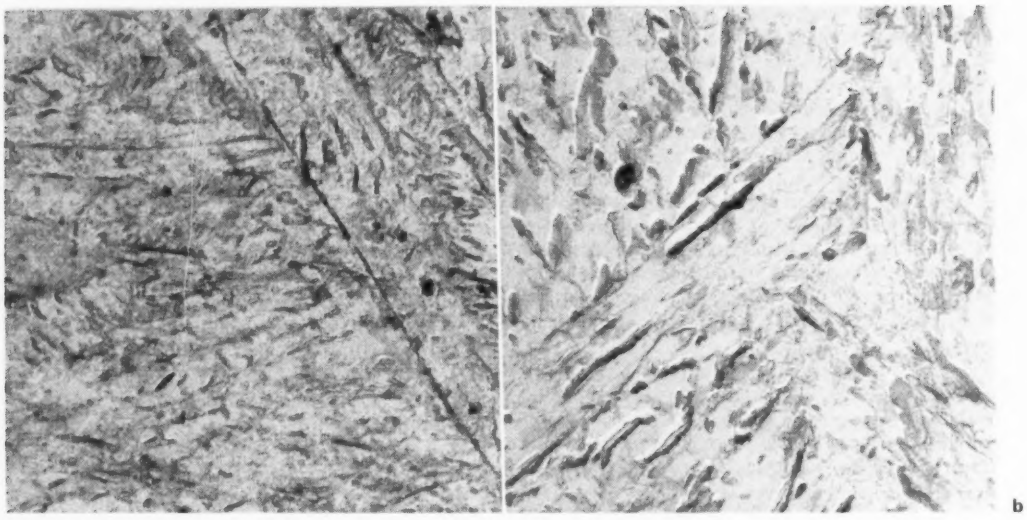
Another major effect of the alloying element is to retard tempering, which has an effect on the structure produced with tempering temperatures between 100° and 300°C. The microstructures of the plain carbon steels shown in Fig.6 after tempering between 100° and 300°C show a considerable precipitation of carbide. A comparable microstructure of a 0·3%C–1%Si steel after tempering at 300°C is shown in Fig.11*a*. There is a quite definite reduction in the amount of carbide precipitation, and this is responsible for the maintenance of higher hardnesses during tempering as shown in Fig.12.

One of the characteristic features of the plain carbon tempered structures was the presence of films of carbide at martensite plate boundaries. It is



a 4%Cr *b* 4%Ni *c* 4%Mn *d* 4%Mo

10 *Effect of alloying elements on the as-quenched structure of* 0·2%C *steel* ×15000



a 0·3%C–1%Si steel; tempered 2 h at 300°C *b* 0·3%C–3%Mo steel; tempered 1 h at 400°C

11 *Effect of* Si *and* Mo *on structure of tempered martensite* ×15000

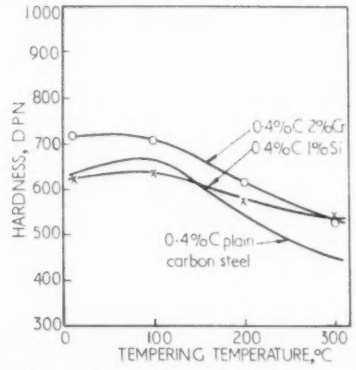
noticeable that the addition of alloying elements modifies this feature, although molybdenum appears to have the most marked effect. The microstructure of a 0·3%C–3%Mo steels is shown in Fig.11*b*. Since molybdenum retards tempering, this steel has been tempered for 1 h at 400°C, to compare with the plain carbon steel tempered at lower temperatures. It will be seen that there has been less precipitation of carbide and there is a greater freedom from carbide films at martensite plate boundaries.

Effect on mechanical properties

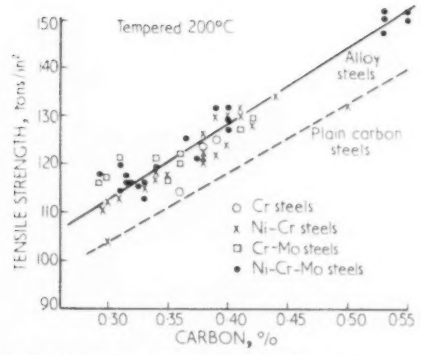
It was concluded from the work on the plain carbon steels that the best properties could be obtained from steels with 0·3–0·5%C using a tempering temperature of about 200°C. Lower-carbon steels suffered from lack of hardenability and higher-carbon steels were susceptible to quench cracking and tended to have retained austenite. It is interesting to see the effect of alloying elements on these lightly tempered martensitic structures. Steels were tested with carbon contents between 0·29 and 0·5% and with many different alloy contents. These were standard commercial alloy steels of the following types: 1%Cr; 1%Cr–0·3%Mo; 3%Cr–½%Mo; 1½%Ni–1¼%Cr; 1½%Ni–1·2%Cr–0·25%Mo. These steels were oil quenched, since they possessed reasonable hardenability and this reduced the danger of quench cracking. They were then tempered at 200°C; the important mechanical properties are summarized in Fig.13. There is a clear relationship between tensile strength and carbon content with all the different alloy steels lying within a relatively narrow band. It will be seen that the alloy steels are at a higher strength level than the plain carbon steels due to their higher initial strength and greater tempering resistance. There is also a relationship between ductility values and tensile strength and once again the alloy steels fall within a general band, although there is a considerable amount of scatter. There were four steels which showed premature failure and low ductility results. These have been shown at the tensile strength recorded, although it is likely that this should have been higher. The general level of ductility of the alloy steels is higher than that of the plain carbon steels, because the results in Table 1 show that even with 0·3%C and a tensile strength level of about 100 tons/in², the elongation values are less than 8% and the reduction of area less than 20%.

With regard to tensile strength and ductility therefore, the alloy steels have better properties than plain carbon steels, but it is not possible to discern any separate alloying effects within the scatter band. There is however, a marked effect of alloying element on the impact properties. In general the impact level decreases with increasing tensile strength, but the alloy steels have considerably better impact properties than the plain carbon steels. Furthermore, the impact values of the steels containing molybdenum all lie above the molybdenum-free Cr and Ni–Cr steels.

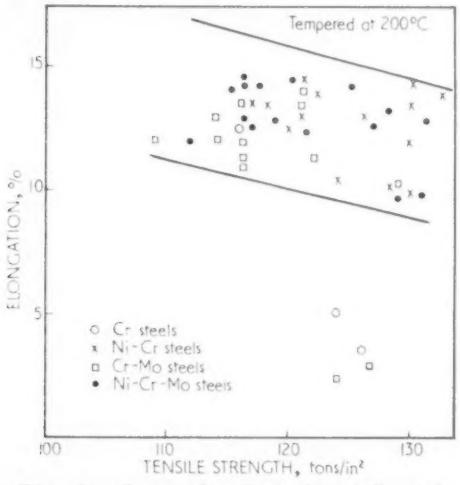
From these results it is possible to relate the mechanical properties to the microstructure. One important effect of alloying elements is that they increase the as-quenched strength and also the tempering resistance. This effect is also visible in the microstructures which show less autotempering in the



12 *Effect of* Cr *and* Si *on tempering resistance of* 0·4%C *steels at low tempering temperature*

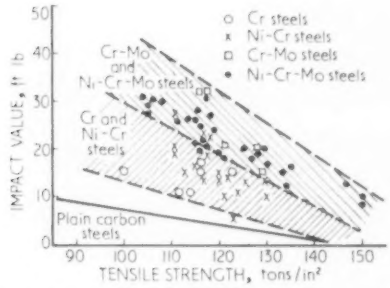


13a *Effect of C content on tensile strength of alloy and plain C steels*

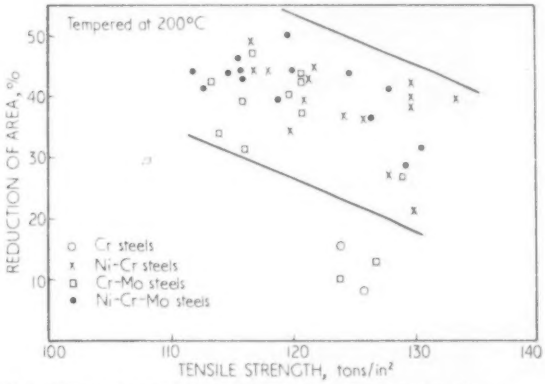


13c *Effect of tensile strength on the ductility of alloy steels*

quen hed structure and retarded carbide formation after tempering. Another important feature of the microstructures is that the alloy steels show fewer carbide films at martensite plate boundaries, particu-

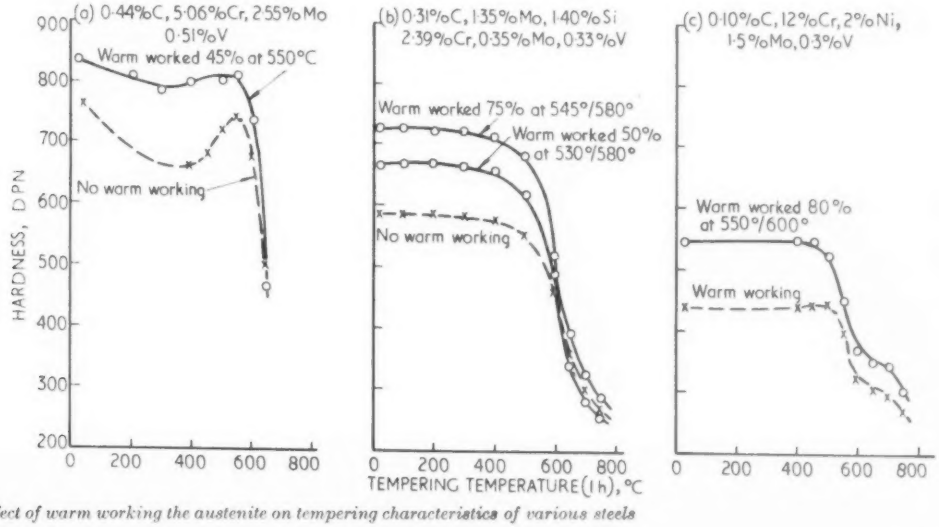


13b *Effect of tensile strength on impact value of plain C and alloy steels showing the beneficial effect of Mo*



13d *Effect of tensile strength on the ductility of alloy steels*

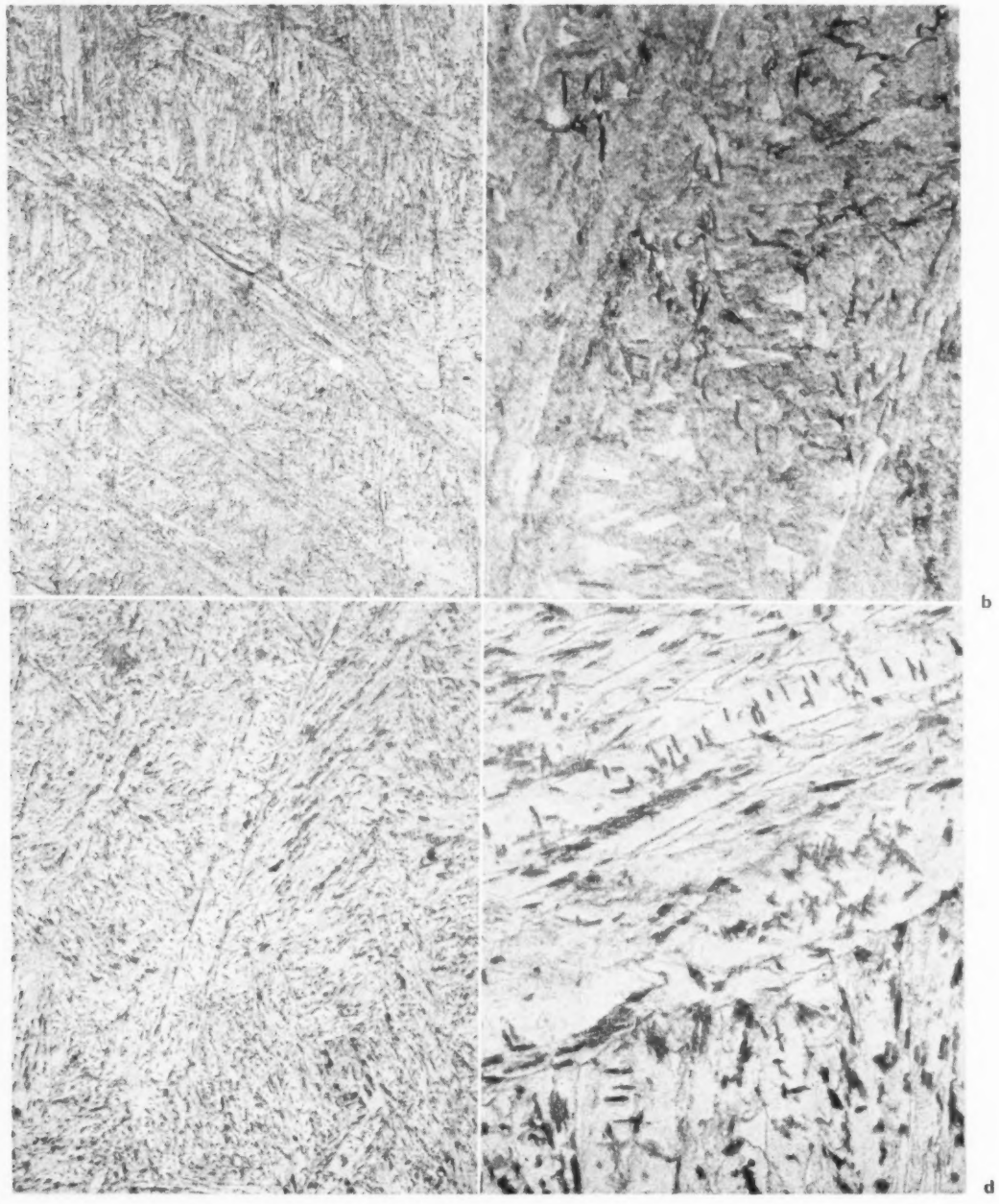
larly in the steels containing molybdenum. It is suggested that this is the main reason why the impact properties are so much better in these steels. It seems clear therefore, that if high tensile properties are required from a steel, the optimum composition is a Ni–Cr–Mo or Cr–Mo with about 0·3%C, and the steel should be quenched and tempered at 200°C. Better ductility can be obtained if the tempering temperature can be raised, but in this case it is difficult to preserve the martensite hardness and a secondary hardening reaction must be used.


(a) 0·44%C, 5·06%Cr, 2·55%Mo 0·51%V
Warm worked 45% at 550°C
No warm working
(b) 0·31%C, 1·35%Mo, 1·40%Si 2·39%Cr, 0·35%Mo, 0·33%V
Warm worked 75% at 545°/580°
Warm worked 50% at 530°/580°
No warm working
(c) 0·10%C, 12%Cr, 2%Ni, 1·5%Mo, 0·3%V
Warm worked 80% at 550°/600°
Warm working
HARDNESS, D.P.N.
TEMPERING TEMPERATURE (1 h), °C

14 *Effect of warm working the austenite on tempering characteristics of various steels*

Effect of warm working

In the preliminary study on plain carbon steels it was suggested that the hardness of martensite was mainly due to a fine grain size and to the solid-solution hardening effect of carbon. While high-tensile strength can be obtained from high carbon steels the ductility values decrease when more than 0·4%C is present. These ductility values are improved when alloy elements are present but even so, poor ductilities are obtained with tensile strengths above 130 tons/in². It seemed therefore, that it was desirable to increase the tensile strength. without increasing the carbon and the only possible way of doing this was to decrease the grain size.



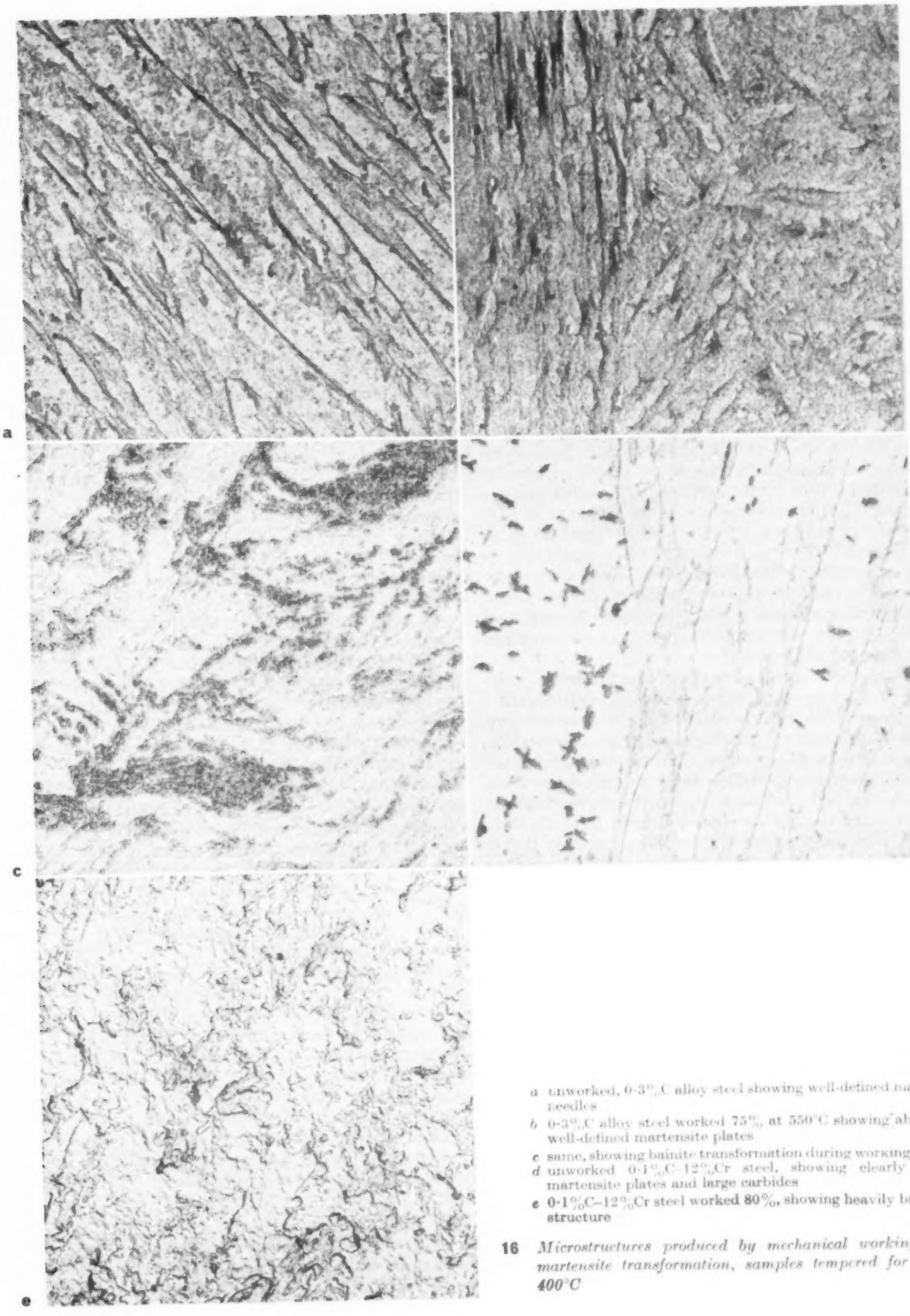
a worked 45% at 550°C, showing heavily deformed regions ×3000

b same, showing small carbide particles and absence of well defined martensite plates ×15000

c unworked material showing well defined martensite plates ×3000

d unworked material showing well developed carbide particles ×15000

15 *Comparison of worked and unworked structures in* 0·44%C, 5%Cr, Mo–V *steel, samples tempered for 1 h at 400°C*



a unworked, 0·3%C alloy steel showing well-defined martensite needles ×15000

b 0·3%C alloy steel worked 75% at 550°C showing absence of well-defined martensite plates ×15000

c same, showing bainite transformation during working ×3000

d unworked 0·1%C–12%Cr steel, showing clearly defined martensite plates and large carbides ×15000

e 0·1%C–12%Cr steel worked 80%, showing heavily broken up structure ×15000

16 *Microstructures produced by mechanical working before martensite transformation, samples tempered for 1 h at 400°C*

It was known that the grain size of martensite is not affected by composition, but there was some evidence to show that the martensitic structure can be modified by austenite deformation.

There has been some published information about this process[5, 6] and it was clear that a steel with a considerably retarded bainite transformation must be used, so that mechanical working can be carried out at a temperature below the recrystallization temperature of austenite, but before any transformation.

This process has been studied on a number of steel compositions. The composition of the first of these was 0·44%C, 5·06%Cr, 0·89%Si, 2·55%Mo, and 0·51%V. This steel was worked by forging at 550°C to produce about 45% reduction in area, before quenching. Further material was available for comparison, in the quenched condition without any working. Hardness values obtained on tempering are compared in Fig.14*a*. This shows a hardness increase of about 150 DPN in the worked specimen and this hardness difference is maintained even after tempering at 400°C. Above 500°C rapid softening takes place in both steels and the hardness difference considerably decreases. Tensile tests were carried out on both steels but unfortunately, premature failures were obtained in both steels at low tempering temperatures. The first satisfactory tensile test was obtained after tempering at 550°C when the worked steel gave 175 tons/in². Comparing this tensile strength with the hardness curves, it is clear that considerably higher strengths would be obtained at low tempering temperatures were it not for premature failures.

The microstructures of these steels clearly show the effect of working on the martensitic structures. The structures which have been selected to illustrate this are in the tempered condition, because the carbide precipitation clearly outlines the structure. However, similar structures existed before tempering. The worked structure showed evidence of deformation and the deformed regions acted as a barrier to the formation of large martensite plates (Figs.15*a* and *b*). This is in marked contrast to the structure of the unworked steel (Figs.15*c* and *d*), which shows quite large martensite plates. In the tempered condition, the carbide particles were considerably more developed in the unworked steel.

Some similar work was carried out on a lower-carbon steel with a composition of 0·31%C, 1·35%Mn, 1·40%Si, 2·39%Cr, 0·35%Mo, and 0·33%V. This steel had a suitable transformation diagram to allow working at about 550°C without transformation and samples were worked at 530–580°C before quenching to produce 50 and 75% reduction of area. The hardness changes on tempering are shown in Fig.14*b* and once again it will be seen that the worked samples have a higher initial hardness than the unworked sample, and that this hardness difference is maintained up to a tempering temperature of about 400°C. The hardness difference with 75% reduction of area was about 150 DPN which was the same as with the previous steel. Tensile tests were also carried out but again low ductility values and premature failures were obtained with the worked samples. After tempering at 200°C, tensile strengths of 148 and 150 tons/in² were obtained from the samples reduced 50% and 75%, compared with 133 tons/in² for the unworked sample. This difference in tensile strength would have been greater if there had not been premature failure in the worked samples and the hardness difference would indicate a tensile strength difference of about 30 tons/in². The microstructures of these steels (Figs.16*a* and *b*) also showed that deformation of the austenite had produced a considerably finer martensite and there was retarded carbide precipitation on tempering. One region of the worked steel showed some bainite which had formed during working (Fig.16*c*), which illustrates that this working process requires close control.

From the work which has been described, it is clear that it is possible to increase the tensile strength of martensite quite considerably by reducing the grain size, but that when 0·3 and 0·4%C steels are used, the inherent ductility is so low that the full strength cannot be realized. It was therefore decided to test a 0·1%C steel and in order to get the necessary hardenability a 12%Cr base composition was selected. This type of steel has the added advantage that the bainite reaction is suppressed, so that it is relatively simple to work the austenite at about 550°C. The composition of the steel used was 0·10%C, 12%Cr, 2%Ni, 1½%Mo, and 0·3%V. Working was carried out at 550–600°C to produce about 80% reduction in area. The hardness changes on tempering worked and unworked material are compared in Fig.14*c* and it is seen that the worked material is about 120 DPN harder and that this hardness difference is maintained up to a tempering temperature of 400°C.

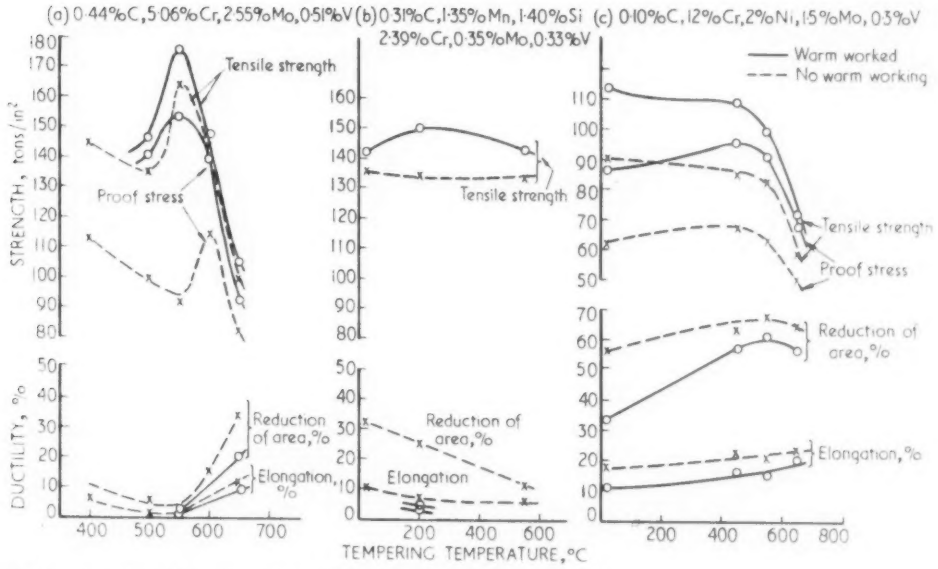
It was possible to obtain a full set of tensile properties from this steel and the results are shown in Fig.17. Up to the tempering temperature of 400°C the worked sample had a tensile strength about 25 tons/in² greater than the unworked sample. Once again, the effect of deformation was clearly visible in the microstructures (Figs.16*d* and *e*). The worked sample has no clearly defined martensitic structure and carbide precipitation has been retarded. From the relationship which is given in Fig.13*a* it can be seen that a 0·1%C steel worked before martensite formation has a tensile strength equivalent to 0·3%C steel in the unworked condition. It is suggested therefore, that there might be considerable interest in the properties of 0·15–0·25%C steels having sufficient alloy content for adequate hardenability after similar deformation before martensite formation.

Further work is now being carried out to assess the effect of warm working before martensite transformation but on the basis of present information there does not seem to be any marked superiority in the combination of mechanical properties which can be obtained from this structure. Although a tensile strength increase of at least 30 tons/in² can be obtained, this can also be achieved by using higher-carbon content in a martensitic structure or by using a secondary hardening process. The authors feel that the latter method is to be preferred.

CONCLUSIONS

From this work it has been possible to see the effect of carbon and alloy content on the structure of martensite and to relate these structures to the properties which can be obtained. The main conclusions are:

1. The major factor controlling the hardness of martensite is the amount of carbon in solid solution. Due to this alone the hardness can be increased from about 200 to 900 DPN.



17 *Effect of warm working the austenite on mechanical properties of various steels*

2. The second most important contribution to the hardness results from the fine grain size. This is responsible for a hardness increment of perhaps 150 DPN compared with the hardness of polygonal ferrite and if a still finer martensite grain size can be produced by deforming the austenite before transformation, then an additional hardness increase of 150 DPN can be obtained irrespective of composition.

3. A third factor is the effect of alloying elements. Due partly to a reduction in autotempering and partly to solid-solution hardening, a hardness increase of 100–150 DPN can be obtained by the addition of up to 4% of alloying element. This effect is greatest in low-carbon steels.

4. The mechanical properties of plain carbon martensites in other than very thin sections are most satisfactory with between 0·3 and 0·5%C, providing that the steels are tempered at 200–300°C. Lower-carbon steels have inadequate hardenability and higher-carbon steels are susceptible to quench cracking and may have retained austenite.

5. When alloying elements are added they have the general effect of raising the initial tensile strength and increasing the tempering resistance. At any tensile strength level the ductility of an alloy steel martensitic structure is higher than a plain carbon steel. Alloy steels also have very much better impact properties, and molybdenum is particularly effective in improving the level of impact energy. This appears to be due to the modification of carbide precipitation so that films at martensite plate boundaries are eliminated.

6. The effect of warm working the austenite before martensite transformation is to raise the tensile strength by at least 25 tons/in². However, on the basis of present evidence, the combination of mechanical properties obtained in this way does not appear to be superior to that which can be obtained either from a higher-carbon martensite or by using secondary hardening.

ACKNOWLEDGMENT

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Activities of manganese oxide in silicate melts

K. P. Abraham, M. W. Davies, and F. D. Richardson

SYNOPSIS

The activities of manganese oxide have been measured at 1500–1650°C in its melts with SiO_2, $CaO+SiO_2$, *and* $CaO+SiO_2+Al_2O_3$. *The results indicate that* MnO *interacts with* SiO_2 *more strongly than* FeO, *but less strongly than* CaO. *In the ternary with* $CaO+SiO_2$ *the pattern of activities is similar to that known for* $FeO+CaO+SiO_2$; *at mole fractions of* SiO_2 *greater than 0·42 mixing of silicates of equal silica mole fraction is nearly ideal; at lower silica concentrations there arise excess integral free energies of mixing, ranging up to −2Kcal per mole of oxide. These can be accounted for in terms of the tendency for such melts to attain a structure in which the manganese tends to associate with* O^{2-} *and* SiO_4^{4-} *ions and the calcium with* $Si_2O_7^{6-}$ *and more complex ions. A semi-empirical method of calculating these excess free energies which was found to hold (within ±200 cal) in the case of* $FeO+CaO+SiO_2$ *melts, holds (within ±400 cal) in this case also. In the quaternary with* $CaO+SiO_2+Al_2O_3$ *at 1650°C, the activity coefficient of* MnO *is constant below 0·07 molar %, but changes greatly as the proportions of the other three components are altered. Manganese losses in the ferromanganese blast-furnace are considered in the light of the results.* 1868

INTRODUCTION

ALTHOUGH the metallurgical importance of silicate melts can hardly be overemphasized, relatively few measurements have been made of the activities of metal oxides in such mixtures. The only firm data so far published are the activities of FeO in its mixtures with silica,[1, 2] and $CaO+SiO_2$,[3, 4, 2] and those of PbO in its mixtures with silica[5] and $CaO+SiO_2$, $MgO+SiO_2$, and $ZnO+SiO_2$. Unfortunately the ranges of composition that can be investigated with the system containing PbO are rather narrow, so the results provide no check on the theories of silicate mixing developed by the authors[6, 7] from the results of Taylor and Chipman for $CaO+FeO+SiO_2$.

With MnO in the region 1500–1650°C, a very large part of its ternary with $CaO+SiO_2$ can be investigated and the mixing theories can thus be put to a reasonable test. Measurements have therefore been made of the activities of MnO in the binary with SiO_2, and the ternary $MnO+CaO+SiO_2$. In view of the importance of slags which contain manganese oxide, from the standpoint of ferromanganese production, measurements have also been made of MnO activities in the quaternary $MnO+CaO+Al_2O_3+SiO_2$ at 1650°C.

The equilibrium which was exploited for these studies was

$$(MnO)=[Mn]+\tfrac{1}{2}O_2 \quad \ldots\ldots(1)$$

The manganese metal was dissolved in solid platinum and the desired oxygen potentials were imposed by means of gas mixtures made from $H_2+CO_2+N_2$. Because this equilibrium has already been studied with pure MnO,[8] the manganese oxide activities can be calculated from the results.

EXPERIMENTAL

Materials

Silica was obtained from clear mineral quartz. This was powdered, washed with dilute hydrochloric acid to remove traces of iron and then with water, and finally dried at 100°C.

Alumina and magnesia were obtained from the Analar oxides which were heated to about 1000°C for 2 h before use.

Lime was prepared from Analar calcium carbonate by calcination at 1000°C for about 2 h.

Manganese oxide was prepared by dissolving electrolytic manganese (99·99%) in concentrated nitric acid and evaporating the solution to dryness. The residue of pure manganese dioxide was baked at 200°C for 1 h and afterwards reduced to green manganous oxide in a stream of hydrogen at 1100°C. Analysis showed the oxide to be virtually stoichiometric.

Platinum Foil 0·002 in. thick (99·98%) was used.

Containers Platinum cups 1 cm dia. and 3 mm deep were made from metal foil 0·005 in. thick. They were supported in trays made from alumina cement, each cup holding about ½ g oxide.

Gas mixtures

Mixtures of H_2, CO_2, and oxygen-free N_2, were used, the proportions being adjusted to give the oxygen potentials required at the temperatures of the experiments. The gases were taken from cylinders and dried by passage through magnesium perchlorate. Mixing was controlled by capillary flow meters containing n-dibutyl phthalate. In some experiments, when the sulphur-holding properties of the slags were also measured, dry SO_2 was metered into the gas stream in addition to the gases mentioned above.

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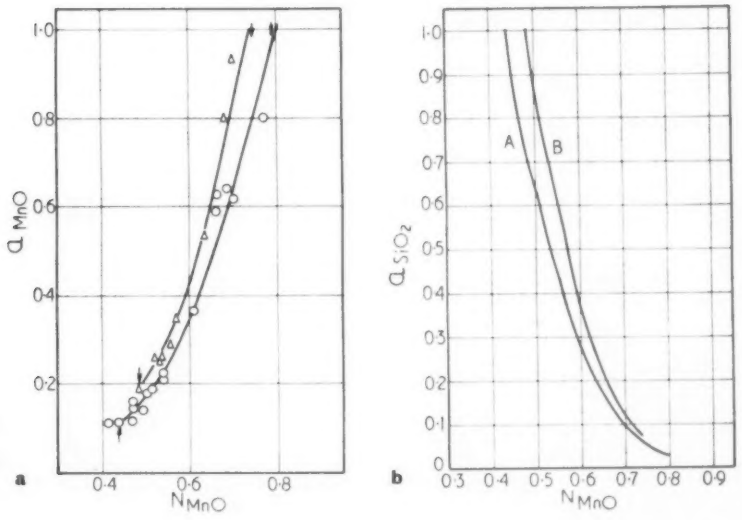


1 *(a) Activities of manganese oxide, relative to solid, in* $MnO+SiO_2$ *melts at 1500°C* △*; and 1650°C* ○. *(b) Activities of silica relative to solid, in* $MnO+SiO_2$ *melts derived from Fig.1a. A at 1500°C, B at 1650°C*

Equilibrium apparatus

This was similar to that used by Fincham and Richardson[9] except for slight modifications. The molybdenum winding of the furnace was supported on a separate alumina tube within which the reaction tube was held. The control thermocouple was inserted between the furnace tube and the reaction tube, the hot junction being at a point about 120°C cooler than the hottest zone. The flowing gases (about 600 ml/STP/min) were made to leave from both ends of the reaction tube, so as to eliminate any lack of control in gas composition, which might arise from the dead space between the end of the gas inlet tube, and the inlet end of the reaction tube.

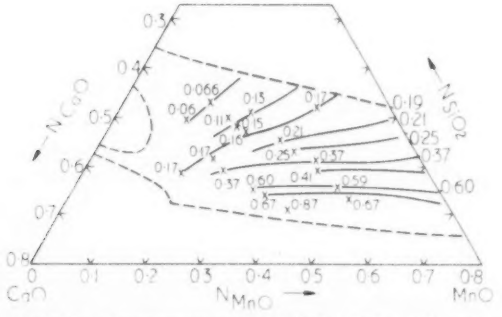
Procedure

Strips of platinum 1·25 in. long and 0·1 in. wide were cut from 0·002-in. foil and wound into open spirals. One was placed in each of the four platinum cups used in an experiment. Appropriate mixtures of the $MnO+SiO_2$ or $MnO+CaO+SiO_2$ were made by grinding the constituents together in an agate mortar. For each experiment, sufficient of one mixture was taken to fill the four cups: in certain cases after the first cup had been filled, a small amount of MnO was mixed into the remaining material, and the second cup filled. The third and fourth cups were filled after similar further additions. The additions were required to make up for the manganese lost from the slags to the platinum spirals and containers during each experiment, and so permit the MnO activity to be obtained for a predetermined slag composition by interpolation or slight extrapolation of the four separate results.

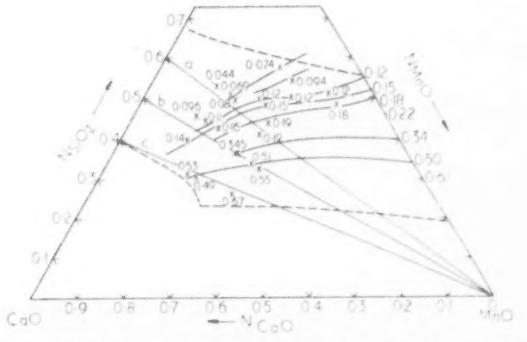
The furnace was brought to the required temperature and the reaction tube flushed with a stream of nitrogen. With the nitrogen still flowing, the tray of cups was placed in the reaction tube and pushed slowly into the central even-temperature zone. The nitrogen was passed for a further 15 min before the controlled gas mixture ($CO_2+H_2+N_2$) was allowed to enter. It was found that equilibrium was established between the melt and the platinum strip with respect to manganese in about 6 h. This was proved by approaching equilibrium from both sides, some spirals being made from pure platinum and others from alloys of manganese and platinum, containing more than the equilibrium amount of manganese.

Analysis

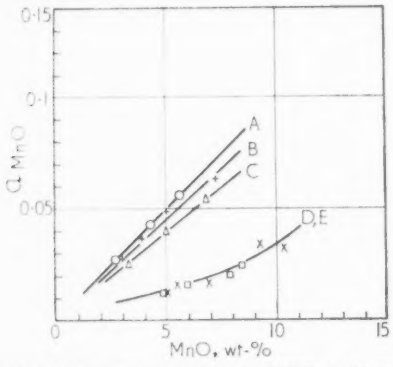
Each oxide sample was dissolved in 5 ml of 40%HF and 15 ml of 50%H_2SO_4. It was heated until fuming started; a little more H_2SO_4 was added, and the whole evaporated almost to dryness. The residue was



2 *Activities of manganese oxide in* $MnO+CaO+SiO_2$ *melts at 1500°C. Broken lines are probable liquidus*

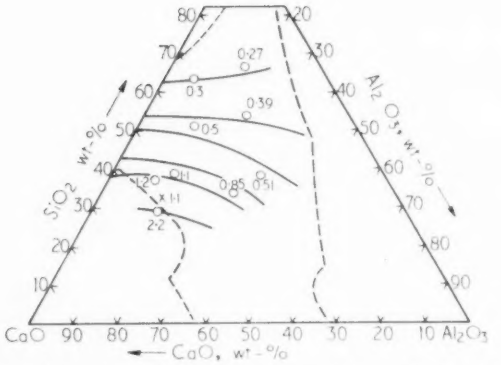


3 *Activities of manganese oxide in* $MnO+CaO+SiO_2$ *melts at 1650°C. Broken lines are probable liquidus*



4 *Activities of manganese oxide at 1650°C in* $MnO+CaO+Al_2O_3+SiO_2$ *melts for various weight ratios of* $CaO:SiO_2:Al_2O_3$

A 52·0: 37·2: 10·8; B 47·2: 38·2: 14·6; C 36·7: 33·3: 30·0; D 30·0: 63·3: 6·7; E 16·7: 66·7: 16·6.



5 *Activity coefficients of manganous oxide (on molar basis) at concentrations up to 0·07 molar (about 8 wt-%) in* $MnO+CaO+SiO_2+Al_2O_3$ *melts at 1650°C. Broken lines show liquidus in the absence of* MnO. *The point marked with a cross is deduced from the results of Filer and Darken*[20] *at 1600°C*

dissolved in water, the manganese oxidized to permanganate with KIO_4, and estimated colorimetrically with a Unicam spectrophotometer.

Each platinum–manganese spiral was dissolved in aqua regia and evaporated to dryness. The residue was treated with concentrated H_2SO_4 and heated until fuming started, so as to drive off HCl. The product was dissolved in water, and the manganese again estimated colorimetrically after being oxidized to permanganate by KIO_4.

RESULTS

The results are shown in Figs.1–5 where activities, or activity coefficients, all relative to the pure solid oxides, are plotted as functions of composition. The activities were calculated from the equation

$$a_{MnO\ (stoichiometric)} = \sqrt{\frac{p_{O_2}}{p^{\circ}_{O_2}}}\, a^{\circ}_{MnO} \quad \ldots\ldots(2)$$

where p_{O_2} is the partial pressure of oxygen prevailing in the experiment; $p^{\circ}_{O_2}$ is the pressure of oxygen which would be in equilibrium with pure MnO and an alloy containing the same amount of manganese as found in the slag experiment; a°_{MnO} is the activity of stoichiometric MnO in the non-stoichiometric solid oxide which would be in equilibrium with $p^{\circ}_{O_2}$. Values for a°_{MnO} and $p^{\circ}_{O_2}$ can be derived from data published elsewhere by Davies and Richardson[10] and Abraham *et al.*[8]

For the binary $MnO+SiO_2$, activities for a number of compositions are listed in Table I. About half the results for the binary were obtained with gas mixtures containing SO_2*as well as $CO_2+H_2+N_2$. There was good agreement between the two sets, so the accuracy of the equilibrium data for the gaseous species COS, SO, S_2, etc., which are not as good as for CO, H_2O, etc., has not seriously affected the calculated activities.

The results for the ternary are more extensive than those published previously[6]; they are slightly different at some compositions but the difference is nowhere greater than 20%. The results for the quaternary are represented in Figs.4 and 5. In Fig.4 the activities of MnO are shown as a function of concentration for a number of ratios of $CaO:Al_2O_3:SiO_2$. It is evident that up to about 8 wt-%, the activity coefficients of MnO are constant at constant ratios of $CaO:Al_2O_3:SiO_2$. It is therefore possible to represent the results by a series of iso-activity-coefficient curves on a ternary diagram as shown in Fig.5, the activity coefficients being based on mole fractions expressed in terms of the component oxides.

* This was added when it was desired to measure the sulphide capacities of the melts at the same time as the MnO activities.

DISCUSSION

Accuracy

The following errors arise in the experiments:

Temperature (reading and calibration of optical pyrometer)	±6°C
Gas composition (calibration and reading of flow-meters)	±1%
Analysis (Mn in slag)	±1%
(Mn in alloy)	±1%
Slag composition (apart from MnO)	negligible

Because the highest pressures of oxygen used in these experiments were 10^{-6} atm, a°_{MnO} never fell below 0·99%, and the correction for the non-stoichiometry of MnO is small compared to the errors of the measurements.

The first three of the errors listed above contribute an error of about ±3% to the derived values of a_{MnO}. To this must be added the error of ±7% present in the relationship between $p^{\circ}_{O_2}$ and wt-% Mn in alloys equilibrated with pure MnO, (*see* Table I of ref.8). Provided that the errors in the constants assumed for the $CO_2+CO+H_2+H_2O$ equilibrium at 1650°C are negligible, the activity date should be correct to about ±8%. The scatter shown in Fig.1*a* (MnO+SiO_2 binary) is consistent with this conclusion.

Binary mixtures

The activities of silica shown in Table I, and Fig.1*b*, have been calculated by application of the Gibbs Duhem equation. The MnO and SiO_2 saturation limits which can be obtained from the recently published phase diagram of Glasser[11] for the MnO+SiO_2 system, are in agreement with the results reported here. The variation of activity with temperature

TABLE I Activities in MnO+SiO$_2$ melts (relative to pure solids)

N MnO	a_{MnO} 1500°C	a_{SiO_2} 1500°C	a_{MnO} 1650°C	a_{SiO_2} 1650°C
0·45	...	...	0·120	0·905
0·50	0·215	0·875	0·175	0·645
0·60	0·425	0·380	0·345	0·290
0·70	0·825	0·115	0·650	0·090
0·74	1·00	0·070	0·780	0·055
0·805	...	...	1·00	0·025

indicates that the partial heat of solution of solid MnO is about 10 Kcal (±5 Kcal) when its mole fraction is 0·5–0·7. This is reasonable compared with the heat of fusion which must be about 9·5 Kcal, but in view of the poor accuracy of the partial heat it is not possible to draw serious conclusions from it.

The values for the activity of MnO in MnO+SiO$_2$ shown in Fig.1*a* are substantially greater than those estimated by Bell, Murad, and Carter[12] from their measurements with mixtures of MnO+FeO+MgO+SiO$_2$ at 1550°C, e.g. about twice as great when N_{MnO} lies between 0·65 and 0·80. In order to get activity values from their measurements of the equilibrium

$$(MnO)+[Fe]=[Mn]+(FeO) \quad \ldots\ldots(3)$$

these workers assumed that γ_{FeO} is unity in the quaternary (it is approximately unity in the binary FeO+SiO$_2$) and that γ_{MnO} in both the systems MnO+FeO+MgO+SiO$_2$, and MnO+SiO$_2$, depends only on the mole fraction of silica. Although Bell *et al.* made measurements of the oxygen present in the metal in equilibrium with their slags, and found this to be approximately proportional to the FeO content, their results are not accurate enough to rule out the possibility that the activity coefficient of FeO is changed by MnO in the ternary mixtures. One would indeed expect that at any constant mole fraction of silica the activity coefficient of FeO would be raised by the presence of MnO and vice versa.[6] Thus from the results of Bell *et al.*, it is only strictly possible to derive values for the ratio a_{MnO}/a_{FeO}.

Now that activity data are available for both the binaries FeO+SiO$_2$, and MnO+SiO$_2$, it is possible to calculate a_{MnO}/a_{FeO} in the slags of Bell *et al.* by application of the ideal mixing laws proposed for silicates by Richardson.[6] It is not possible to calculate the MnO activities, because substantial amounts of magnesia dissolved from the crucibles and accurate activities are not available for the binary MgO+SiO$_2$. The activity ratios are given by the equation,

$$\frac{a_{MnO(q)}}{a_{FeO(q)}}=\frac{N_{MnO(q)}\,a_{MnO(b)}}{N_{FeO(q)}\,a_{FeO(b)}}\sqrt[x]{\frac{a_{SiO_2(b\cdot Mn)}}{a_{SiO_2(b\cdot Fe)}}} \quad \ldots\ldots(4)$$

where q and b indicate quaternary and binary respectively, and the mole fraction of silica in both is equal to $1/(1+x)$ i.e. the melt formula is $xMO.SiO_2$.

Values of $a_{MnO}(q)/a_{FeO}(q)$ derived from equation (4) by application of Schuhmann and Ensio's[1] activity data for FeO+SiO$_2$, and the authors' results for MnO+SiO$_2$ are shown in Table II where they are compared with the main results of Bell *et al.* It is evident that there is no serious conflict between the two sets of experimental results provided the quaternary data are treated in this way. Indeed, the results of Bell *et al.* lend support to the mixing theory.

The equilibrium represented by equation (3) has

TABLE II Activity relationships in MnO+FeO+MgO+SiO$_2$ melts at 1550°C

N_{MgO}	N_{MnO}	N_{FeO}	N_{SiO}	$a_{MnO(t)}/a_{FeO(t)}$ Experimental	$a_{MnO(t)}/a_{FeO(t)}$ Calculated
0·079	0·510	0·255	0·156	1·52	1·33
0·082	0·392	0·385	0·141	0·77	0·66
0·156	0·277	0·353	0·212	0·37	0·46
0·101	0·413	0·305	0·182	0·84	0·85

Standard states for MnO and FeO both liquid. To convert the authors' values for the activities of solid MnO in MnO+SiO$_2$ melts to values for liquid MnO, an entropy of fusion of 5 cal deg.$^{-1}$ (as for FeO) has been assumed, the melting point being 1850°C.

been investigated by Korber and Oelsen[13] at 1450° and 1550°C with simple FeO+MnO melts and FeO+MnO+SiO$_2$ slags saturated with silica. In the latter mixtures, the compositions ranged from 50 wt-%MnO +50 wt-%SiO$_2$ to 48 wt-%FeO+52 wt-%SiO$_2$. These investigators found that an equilibrium constant expressed in wts-% and of the form

$$K=\frac{(MnO)}{(FeO)}\cdot\frac{100}{[Mn]},$$

held for both sets of melts, but in the ternary mixtures saturated with silica, it was six times greater than in simple FeO+MnO melts. In the latter, one can safely assume that the activity coefficients of MnO and FeO (relative to the liquid state) are both unity, so that at silica saturation

$$\frac{\gamma_{MnO(t)}}{\gamma_{FeO(t)}}=\frac{1}{6},$$

where (t) indicates ternary.

If equation (4) holds, one can write

$$\frac{a_{MnO(t)}}{a_{FeO(t)}}=\frac{N_{MnO(t)}}{N_{FeO(t)}}\cdot\frac{a_{MnO(b)}}{a_{FeO(b)}}\sqrt[x]{\frac{1}{1}},$$

where (b) indicates binary MnO+SiO$_2$, or FeO+SiO$_2$ and x has a value of 0·82±0·03—the silica content at silica saturation increasing by 2 wt-% silica as the mixtures change from MnO+SiO$_2$ to FeO+SiO$_2$. It then follows that a silica saturation

$$\frac{a_{MnO(b)}}{a_{FeO(b)}}=\frac{1}{6}.$$

From the results of Schuhmann and Ensio[1] $a_{FeO}(b)$ is about 0·4, so that $a_{MnO}(b)$ (relative to liq. MnO) is 0·07. Relative to the solid at 1550°C, the value would be 0·1, and this is reasonably close to the value that can be interpolated for 1550°C from Fig.1*a*.

Attempts have been made by Turkdogan and Hancock[14] to deduce silica activities for the MnO+SiO$_2$ mixtures from measurements of the equilibrium,

$$(SiO_2)+2C(\text{graphite})=[Si]+2CO,$$

where the silicon is dissolved in liquid manganese saturated with carbon. Unfortunately our knowledge of the thermodynamics of melts containing manganese, silicon, and carbon is not sufficient to permit such deductions,[15] so it is more to the point to couple their equilibrium results with the activities of manganese oxide reported here, in order to derive activities in the metal solutions. It is necessary to neglect those experiments with slags containing more silica than the

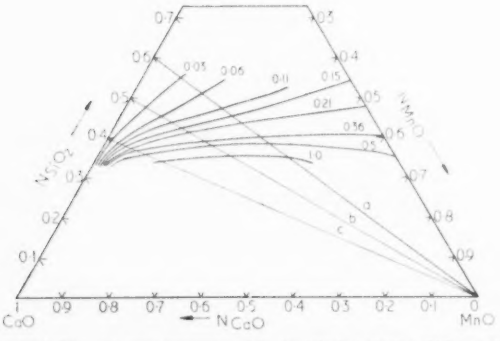
TABLE III Manganese and iron activity coefficients in melts containing silicon and saturated with graphite

Mole fractions			Activity coefficients	
			Mn at	Mn at
Mn	Si	C	1400°C	1450°C
0·71	0·08	0·21	...	0·62
0·71	0·10	0·19	...	0·70
0·71	0·13	0·16	...	0·54
0·72	0·01	0·27	0·74	...
0·72	0·03	0·25	0·60	...
0·72	0·10	0·19	0·30(?)	...
Fe	Si	C	Fe at 1420°C	
0·79	0·05	0·16	0·80	
0·75	0·15	0·10	0·78	

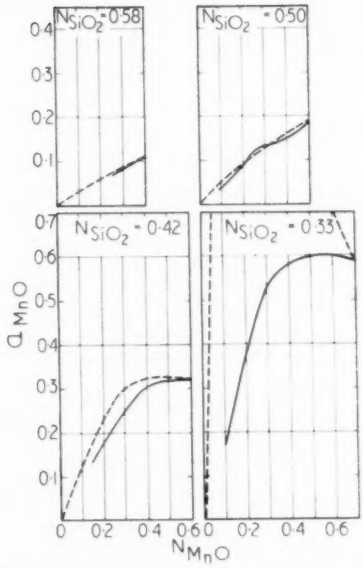
phase diagram[11] would allow and to recognize that the accuracy of the results in terms of N_{Si} in the metal, is no better than $\pm 0{\cdot}02$. If this is done, the results shown in the upper part of Table III are obtained. For the purposes of comparison typical values for γ_{Fe} in corresponding iron melts are shown in the lower part of the table. These have been taken from the results of Chipman *et al.*[16] Although the manganese melts contain more carbon, the activity coefficients are nearly the same as for iron.

Ternary MnO+CaO+SiO$_2$

The ternary results can be used to test the theory of silicate mixing to which reference has already been made. The activity curves which may be calculated from activities in the two binaries MnO+SiO$_2$ and CaO+SiO$_2$[17, 6] are shown in Fig.6. This may be compared with Fig.3, but a better comparison between the experimental and calculated activities can be obtained from Fig.7. Agreement is good when N_{SiO_2} exceeds 0·42. The significance of the differences at lower concentrations of silica can best be appreciated from Fig.8 which shows the free energies of formation, from the crystalline oxides, of CaO+MnO+SiO$_2$ melts at 1600°C. The full lines (*a*), (*b*), and (*c*), have been calculated from the binary free energy curves shown in Fig.9, for compositions that lie along the lines (*a*), (*b*), and (*c*) of Fig.6. The broken lines show the free energies of formation which would be compatible with the measured MnO activities. It appears that the excess free energies (i.e. the deviations from ideal silicate mixing) are a maximum when N_{SiO_2} is about 0·3, and they do not exceed -2 Kcal mole^{-1} in the worst case.*

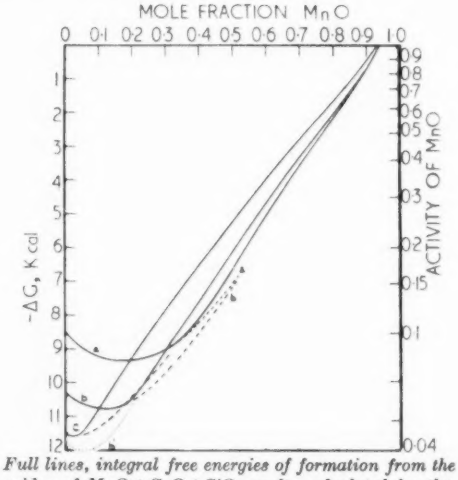


6 *Activities of manganous oxide in* MnO+CaO+SiO$_2$ *melts at 1600°C calculated for ideal mixing of silicates from activities in the binaries* CaO+SiO$_2$ *and* MnO+SiO$_2$

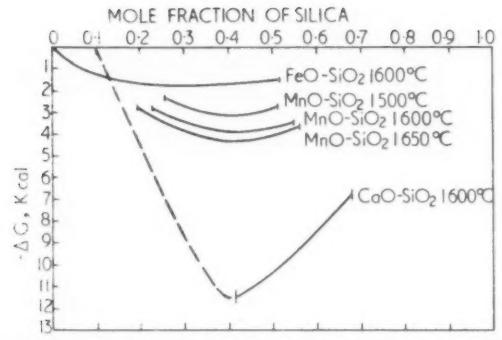


7 *Manganous oxide activities in* MnO+CaO+SiO$_2$ *melts at 1600°C. Full lines experimental results: broken lines calculated values. The inflexion in the curve for the metasilicate mixtures occurs both at 1650° and 1500°C and is outside the range of standard deviations*

* If the data of Chipman[16] are used for the binary CaO+SiO$_2$, the points from which the ΔG curves start at the left-hand side of Fig.8, would be lowered (note the lightly dotted curve b′ which would take the place of b) and the deviations would be rather greater. In view, however, of other supporting data,[23, 24] the authors prefer the data for CaO+SiO$_2$ on which the full lines are based.



8 *Full lines, integral free energies of formation from the solid oxides of* MnO+CaO+SiO$_2$ *melts calculated by the ideal silicate method.*[6] a, b, *and* c *indicate that the compositions lie along the lines* a, b, *and* c *in Fig.7. Broken lines show free energy curves compatible with the measured* MnO *activities. Lightly dotted line* b′ *shows the effect of using Chipman's data on* CaO+SiO$_2$ *in calculations for melts lying along line* b *in Fig.7*



9 *Integral free energies of formation of binary silicate melts:* $FeO+SiO_2$[1]; $MnO+SiO_2$ *this paper*; $CaO+SiO_2$ [17,6]

where the excess negative free energies are somewhat greater. Although the information at present available on silicate melts is insufficient to permit the development of a theory that fully accounts for these excess values, certain general conclusions can be drawn, and the values can be calculated to within 400 cal by a semi-empirical method.

It now seems clear that, in any silicate melt, there must be a distribution of anionic groups ranging from $O^=$ and SiO_4^{4-} through $Si_2O_7^{6-}$ to substantial groups such as $Si_6O_{15}^{6-}$. Chromatographic studies on water soluble phosphate glasses[19] have shown that quite a wide distribution of groups may be possible: the fact that the heats of formation of the silicate melts from the component oxides are much smaller than those of phosphate glasses, suggests that the distribution will be wider with the silicates than has been observed with the phosphates. It is evident, therefore, that the basic premise of the theory of ideal mixing of silicate melts is not strictly correct, because the distribution of anionic groups for one metal silicate will be different from that for another, e.g. the distribution will be wider for xFeO:SiO_2 than for xCaO.SiO_2. However the contribution to the free energy of formation of a ternary melt arising from the entropy of mixing these differently distributed anionic groups, would be substantially less than that arising from mixing the two different cations: within the accuracy of the data so far available for such mixtures this anionic configurational entropy can reasonably be neglected.

For mixtures in which N_{SiO2} is less than 0·4, there is the further possibility that substantial changes in anionic structure occur when two binary melts of equal silica mole fraction are mixed to make a ternary. It has been shown,[7] for example, that in the system $FeO+CaO+SiO_2$, a melt consisting of 0·30 moles of CaO, 0·20 of SiO_2, and 0·5 of FeO, would have a lower free energy when split into two liquids (equation (5)) than if it had the structure resulting from ideal mixing of the cations about the silicate ions present in the separate iron and calcium binaries (equation (6)).

$$(0{\cdot}30\ CaO+0{\cdot}075SiO_2)+(0{\cdot}50FeO+0{\cdot}125SiO_2)= (0{\cdot}30CaO+0{\cdot}20SiO_2)+(0{\cdot}50FeO) \quad (5)$$

ΔG (2 liquids) = −3·0 Kcal.

$$(0{\cdot}30CaO+0{\cdot}075SiO_2)+(0{\cdot}50FeO+0{\cdot}125SiO_2)= \text{random single melt} \quad (6)$$

ΔG (ideal silicate) = −2·0† Kcal.

Thus it seems probable that the single solution that is actually produced on mixing, has a structure in which there is a higher proportion of oxygen ions and $Si_2O_7^{6-}$ ions than in either binary. One would expect such a solution to have a distribution of cations which is non-random and reflects the tendency to form two separate liquids; the iron ions tending to associate with the oxygen ions, and the calcium with the higher silicate ions. The free energy of formation of such a solution would, therefore, be equal to ΔG (2 liquids) in equation (5), plus a free energy of mixing, substantially less than that which would arise from athermal random mixing of the ions of the two liquids.

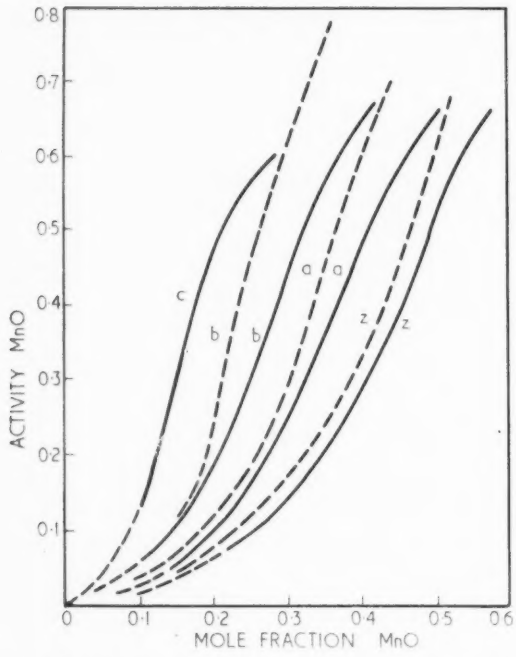
TABLE IV Results of free energy

Melt composition in moles			ΔG of formation in Kcal/mole of oxide			
SiO_2	CaO	MnO	Ideal silicate	2 Liquids	Calculated	Actual ±0·3
0·33	0·50	0·17	−9·7	−9·5	−10·4	−10·7
0·30	0·45	0·25	−8·6	−8·4	−9·5	−9·8
0·30	0·30	0·40	−7·5	−7·0	−8·1	−8·5
0·30	0·20	0·50	−6·6	−5·9	−6·8	−7·0
0·28	0·42	0·30	−7·9	−7·8	−8·9	−9·3
SiO_2	CaO	FeO				
0·33	0·50	0·17	−9·8	−9·6	−10·5	−10·5
0·30	0·45	0·25	−8·0	−8·6	−9·7	−9·8
0·30	0·20	0·50	−5·5	−5·0	−5·9	−6·1

A semi-empirical method of calculating the free energy of formation of ternary melts in which N_{SiO2} is less than 0·40 has already been proposed[7] and applied to the system $FeO+CaO+SiO_2$: it gives rise to the integral excess free energies which are correct to within ±200 cal. It is interesting to see how far this calculation* is successful in deriving the corresponding free energies for the system $MnO+CaO+SiO_2$. The figures are shown in the upper half of Table IV; in the lower half some of the figures previously published for $FeO+CaO+SiO_2$, are shown for comparison. ΔG (ideal silicate) is the free energy of formation per mole of oxide, calculated from the ideal silicate mixing law,[6] ΔG (2 liquids) is that for the formation of two liquids with this same overall composition, one liquid having the composition $3CaO.2SiO_2$. It can be seen that although the integral free energy curves for $MnO+SiO_2$ are substantially different from those for $FeO+SiO_2$ (*see* Fig.9), this method of calculation holds reasonably well although it may be consistently low. The difference, however, is hardly significant. Although the calculation leaves much to be desired, it appears to be applicable over a reasonable range of conditions, and to support the general contention that the excess negative free energies of mixing arise in the main,

† This figure takes account only of the entropy of mixing the cations. Mixing of the different anion distributions is not likely to add more than 300 cal. to ΔG, compared with five or more times this amount for the cations.

* To make the calculation, first calculate ΔG for the formation of 2 liquids from the binary ΔG curves: one liquid having the composition $3CaO.2SiO_2$ and the other consisting of the remaining MnO and SiO_2, or MnO and CaO. This is ΔG (2 liquids) in Table IV. To this add 40% of the configurational free energy of mixing which would in theory be obtained if the liquids consisted solely of $Si_2O_6^{6-}$, SiO_4^{4-}, and O^{2-} anions and Ca^{2+} and Mn^{2+} cations and these were mixed randomly and athermally irrespective of size, cations with cations and anions with anions.



10 *Activities of* MnO *relative to solid in the system* $MnO+CaO+SiO_2$ *at various ratios of lime to silica. Full lines 1650°C; broken lines 1500°C;* $CaO:SiO_2$*; mole ratios* (a) 4:6, (b) 5:5, (c) 6:4, (z) 3:7

because the anionic structures of the ternary melts are different from those of the binaries when N_{SiO_2} is less than 0·4.

From the data considered earlier for the ternary $MnO+FeO+SiO_2$ (Table II), it appears that even when N_{SiO_2} is less than 0·40, this system follows the ideal silicate mixing equations fairly closely. Because the ΔG curves for $FeO+SiO_2$ and $MnO+SiO_2$ lie much closer together than for $MnO+SiO_2$ and $CaO+SiO_2$, and because the curve for $MnO+SiO_2$ is much less steep than for $CaO+SiO_2$, the excess free energies of formation which would arise in the ternary $MnO+FeO+SiO_2$ from changes in the anionic structure, turn out to be positive. Thus one would expect ideal silicate mixing to occur because, in this case, it leads to the lower free energy.

Quaternary $MnO+CaO+SiO_2+Al_2O_3$

The activity coefficient contours in Fig.5 were drawn to the $CaO+SiO_2$ limit, by estimating MnO activity coefficients at low concentrations in $CaO+SiO_2$ melts from the plots of a_{Mn} versus N_{MnO}, shown in Fig.10 for compositions lying along the lines (*a*, *b*, *c*) etc. in Fig.3. The results may be compared with certain existing data. Filer and Darken[20] have brought samples of iron blast-furnace slag and metal into equilibrium in the presence of carbon, and have measured the partition of manganese. About 0·2 wt-% of manganese oxide was present in their slags which contained between 2 and 6% of MgO besides $CaO+SiO_2+Al_2O_3$. The MnO activity coefficient can be calculated, if it is assumed that γ_{Mn} is about equal to γ_{Fe} in carbon-saturated metal (*see* Table III). For the slag

TABLE V Slag–metal relationships for ferromanganese equilibria

Temp., °C	a_{Mn}	a_{MnO}	γ_{MnO}	N_{MnO}	MnO in slag, Wt-%	Si in metal, approx., Wt-%
1500	0·5	0·17	1·3	0·14	15	1·5
1550	0·5	0·09	1·3	0·08	8	3
1600	0·4	0·05	1·2	0·04	5	8
1700	0·4	0·01	1·2	0·01	1	14

γ_{MnO} from Fig.5; a_{Mn} from Table III; %Si calculated on assumption γ_{Si} in ferromanganese the same as in iron.

used in their tests at 1600°C, the value of γ_{MnO} is 1·1 compared with about 1·4 interpolated from Fig.5.

A blast-furnace ferromanganese slag may typically have the composition in wt-%:

MnO	10	SiO_2	35
Mn metal	2	Al_2O_3	11
Fe + FeO	0·2 to 0·5	CaO, MgO, BaO	40
S	2		

This is produced in contact with a metal containing some 80%Mn, 7%C, and 1·5%Si and its temperature is about 1550°C. If equilibrium were attained between slag and metal in the presence of carbon and CO at 1 atm one can calculate the figures in Table V from the available thermodynamic data, assuming that the activity coefficient of MnO rises slightly as the temperature falls.

In ferromanganese practice one is thus fairly close to equilibrium with respect to manganese but fortunately further removed from equilibrium with respect to silicon. This is precisely what has been observed by Filer and Darken[20] with the iron blast-furnace. The data in Table V underline the impossibility of attaining better recoveries of manganese in the presence of such slags by going to higher temperature without obtaining more silicon in the metal. For better recoveries one must turn to more basic slags in which the activity coefficient of MnO is higher while that of SiO_2 is lower.

Both the values and the contours shown in Fig.5 differ profoundly from those derived by Turkdogan,[21] from the results of Stuckel and Cocubinsky.[22] No explanation can be offered. It can only be said that the very low values given by Turkdogan for γMnO are quite out of line with the results of Filer and Darken[20] and with slag–metal relationships in ferromanganese blast-furnace practice.

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Correspondence on the paper

A new concept of hydrogen embrittlement in steel*

J. G. Morlet, H. H. Johnson, and A. R. Troiano

Professor **P. G. Bastien,** Monsieur **P. Y. R. Azou,** and Monsieur **J. P. Plusquellec,** of the Centre de Recherches de l'Ecole Centrale des Arts et Manufactures, Paris, wrote: We are pursuing research on a neighbouring subject at present, and we think that we can contribute something more to the understanding of the problem.

In 1951,[1] two of us put forward a hypothesis regarding the process of embrittlement of steel by hydrogen, and we think it useful to recall that theory before going more deeply into the discussion.

Hydrogen contained in metal can be situated in different places:

(i) either in the perfect parts of the cubic-centred lattice, or
(ii) in imperfections; (*a*) minor: vacancies and dislocations; (*b*) medium: dislocation aggregates; (*c*) major: voids such as microcracks, grain boundaries, blowholes, etc.

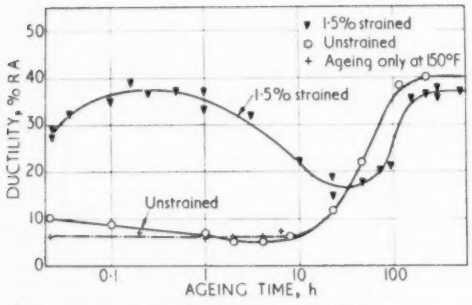
We think that, in the absence of external signs during elastic deformation and the beginning of the plastic phase[2] the hydrogen causing brittleness is situated in the perfect parts of the lattice and in the minor imperfections, in relation, as we have shown, with the (112) family of planes.[3] Hydrogen localized in that way in the form of protons would be the only factor responsible for the lowering of ductility, in so far as the quantity of gas absorbed would not be enough to lead to macroscopic deteriorations in the medium imperfections; in the latter case it would be present in atomic or even molecular form. This last state, however, is certainly less mobile than that of the proton, a dissociation having necessarily to precede a diffusion in ionic form.

Brittleness, as well as the way in which it is affected by the temperature of test and speed of deformation,[4] is explained by the temporary afflux of the protons brought by the dislocations piling up in front of an obstacle and producing, as is well known, a microcrack ahead of the pile-up. The protons conveyed in that way come either from Cottrell atmospheres or from perfect parts of the lattice. Embrittlement thus requires a work-hardening which causes, in the microcracks considered above, an accumulation of hydrogen corresponding to a concentration higher than that of the state of statistical equilibrium. Our theory is thus in no way equivalent to that of Zappfe, which assumes a static pressure. It is based on a dynamic drift of hydrogen giving rise to the occurrence of temporary stresses around the microcracks produced during the deformation. We particularly emphasize the fact that the simultaneous occurrence of two factors is necessary: the presence of hydrogen having sufficient mobility, and production by work-hardening of microcracks ahead of the dislocation pile-ups.

The experimental results of Morlet *et al.* can be explained on the basis of this theory. These authors, working with medium-carbon steel, subjected tensile test pieces to the following treatment: loading in hydrogen without permanent deformation; work-hardening in liquid nitrogen; holding at different temperatures (between −30°C and +150°C) for various times; breaking at ambient temperature. The curve in Fig.*A* summarizes the results obtained. It thus appears that a work-hardening in liquid nitrogen brings about a decrease of brittleness when the tensile test is carried out at ambient temperature. This can easily be understood. At the temperature of liquid nitrogen the protons of the minor imperfections are incapable of following the displacement of the latter which may take place in slip or twinning processes. As a result, therefore, these protons will be found in supersaturations in parts of the lattice which have become perfect again owing to the mechanical diffusion of the imperfections. This condition will be maintained as long as the mobility of the protons is not increased; in other words, as long as the temperature is not raised again. Consequently the dislocations will no longer be accompanied by their Cottrell cloud and the total brittleness will be temporarily decreased. This effect will be more marked in proportion to the work-hardening in liquid nitrogen, i.e. the greater the number of imperfections displaced at that temperature.

These local supersaturations will be progressively reabsorbed by a thermal diffusion of the protons towards the new positions of the imperfections. The brittleness would be bound, therefore, to have a tendency to increase by the formation of new atmospheres. However, during the first period of holding at ambient temperature a contrary phenomenon can occur. For if certain points are supersaturated, a dislocation passing through them will reconstitute its atmosphere completely. Thus a possibility of embrittlement arises, a possibility still greater because of the particular properties of

* J. G. Morlet *et al.*: *JISI*, 1958, **189,** 37–44.
The correspondence on this paper began on 28 November 1958.



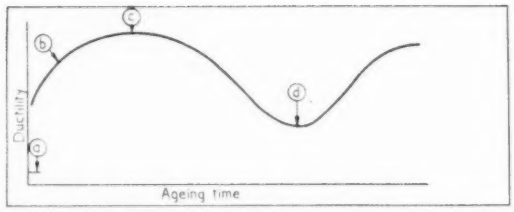
A *Summary of results obtained by Morlet* et al.

certain planes of the lattice in regard to plastic slip and localization of hydrogen. With thermal diffusion of hydrogen away from these points of supersaturation, then, one possibility of embrittlement disappears. While holding at ambient temperature after work-hardening in liquid nitrogen, therefore, two opposing processes come into play:

(i) embrittlement increasing with respect to time by reformation of proton atmospheres round minor imperfections
(ii) ductility of low order, increasing with respect to time by disappearance of points of supersaturation.

Thus, in our opinion, the different phases of the curve (Fig. *B*) of Morlet *et al.* are explained, the ductility maximum being the result of the two processes (i) and (ii). It will be noticed that the lower the holding temperature, the slower are the diffusion phenomena. It is thus normal that the time corresponding to the ductility maximum should be displaced towards greater values when the holding temperature decreases.

It follows that at the end of a long holding time the values of ductility obtained for steels, whether work-hardened or not, must tend to come closer together. Moreover, in view of the general evolution of hydrogen towards the outside of the lattice, an overall diminution of brittleness must take place, a phenomenon which will occur more rapidly as hydrogen diffuses more freely towards the outside, and hence as work-hardening



B *Different phases in curve from results of Morlet* et al.

in liquid nitrogen, which is the source of minor and medium imperfections, becomes less important.

AUTHOR'S REPLY

Professor **A. R. Troiano** (Case Institute of Technology, Cleveland, Ohio, USA) wrote in reply: We wish to thank Messrs Bastien, Azou, and Plusquellec for their interesting comments. We feel that the discussion as presented here gives a somewhat modified picture of the discussers' original theory and may now be more appropriate to a possible explanation of the phenomena described in this paper. This depends largely on what is meant by 'a dynamic drift of hydrogen giving rise to the occurrence of temporary stresses around the microcracks produced during the deformation'. If gaseous hydrogen (molecular) is inferred then the picture is not much different than before; but if hydrogen protons are meant, then the theory is substantially different than was originally presented.

Shortly, we expect to publish in the *Revue de Métallurgie* a further explanation of the phenomena presented here, in terms of a lowering of the fracture stresses by hydrogen in front of the piled-up dislocation array. This increase in energy and lowering of the fracture stress is believed to be brought about by the increased repulsive forces of overlapping electron d-bands.

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STAL *in English*

No 6 of the 1960 series of *Stal in English* has been published and subsequent issues will appear at about three-weekly intervals. Full details of subscriptions to *Stal in English* are available from the Secretary of The Iron and Steel Institute.

NEWS | *Announcements and News of Science and Industry*

THE IRON AND STEEL INSTITUTE

AUTUMN GENERAL MEETING 1960

As previously announced, the Autumn General Meeting of the Institute will be held at Church House, Great Smith Street, London SW1, on Tuesday and Wednesday, **29 and 30 November 1960,** beginning at 9.30 am.

The proceedings will include a meeting on fluid flow in furnaces and converters (*29 and 30 November*), and a meeting on martensite (*29 November*). Full details of these meetings were given in the July 'News' (p.341).

SYMPOSIUM ON STEELS FOR REACTOR PRESSURE CIRCUITS

The Institute is also arranging a symposium on steels for reactor pressure circuits. This will be held at the Hoare Memorial Hall, Church House, Great Smith Street, London SW1 from Wednesday to Friday, **30 November to 2 December 1960.** It will be included in the activities of the British Nuclear Energy Conference, of which the Institute is a member.

The provisional programme is as follows:

Wednesday, 30 November

9.30–10.30

Opening session and introductory lecture by Sir Leonard Owen, C.B.E., M.ENG. (member for and managing director of the Production Group, UK Atomic Energy Authority).

10.30–12.45 Session I High-temperature properties

Presentation and discussion of:

'The high mechanical properties of the steels used in gas-cooled reactor pressure vessels' by J. M. Robertson (C. A. Parsons Ltd) and R. W. Nichols (UKAEA).

'Metallurgical aspects of creep' by J. Glen (Colvilles Ltd) and J. D. Murray (United Steel Companies Ltd).

'The subcritical graphitization of steels and related alloys' by J. E. Harris, J. A. Whiteman, and A. G. Quarrell (Sheffield University).

2.45–5.30 Session II Corrosion

Presentation and discussion of:

'The oxidation of reactor steels in carbon dioxide' by L. E. Benson *et al.* (AEI (Manchester) Ltd)

'The corrosion of pressure circuit materials in boiling and pressurized water reactors' by F. L. LaQue and M. A. Cordovi (International Nickel Co. Inc.)

'The corrosion of steels in liquid sodium' by C. Tyzack (UKAEA).

Thursday, 1 December

9.30–11.30 Session III Fabrication aspects

Presentation and discussion of:

'Metallurgical problems associated with engineering and fabrication of gas-cooled reactor pressure vessels' by W. C. Holliday and M. J. Noone (Whessoe Ltd)

'Fabrication of the reactor vessel heat exchanger and circuitry for the Dounreay fast reactor' by P. K. Richards (John Thompson Water Tube Boilers Ltd)

(*It is hoped also to include a paper on fabrication in water-cooled reactors*)

11.30–12.45 and *2.45–5.00*

Session IV Irradiation effects

Presentation and discussion of:

'Theoretical aspects on irradiation damage and brittle fracture in steel pressure vessels' by A. H. Cottrell (Cambridge University)

'The effects of neutron irradiation on the ductile–brittle transition temperature of some pressure-vessel steels' by D. R. Harries (AERE), C. Judge (Admiralty Materials Laboratory), and R. W. Nichols (UKAEA)

'Practical neutron dosimetry for steel irradiation studies' by L. E. Steele (US Naval Research Laboratory)

'Studies of radiation effects and recovery of notch ductility of pressure-vessel steels' by J. R. Hawthorne (US Naval Research Laboratory)

'Mechanical properties of irradiated steels' by R. G. Berggren (Oak Ridge National Laboratories)

'Influence of some metallurgical variables on irradiation embrittlement of ferritic steel' by L. P. Trudeau (International Nickel Company of Canada Ltd).

Friday, 2 December

9.30–12.45 and *2.45–4.30*

Session V Steels for future reactors

Presentation and discussion of:

'Progress in development of some improved steels for reactor pressure vessels' by I. M. Mackenzie (Colvilles Ltd)

'The development of a high-tensile plate steel for use at elevated temperatures and pressures' by T. F. Pearson and M. Z. de Lippa (Consett Iron Co. Ltd).

'Steels for nuclear power plant' by K. J. Irvine and J. D. Murray (United Steel Companies Ltd)

'The effects of niobium on the properties of low-alloy steels' by J. E. Russell (English Steel Corporation).

The papers for the symposium are to be preprinted and will appear in due course, together with reports of the discussions, in the Institute's Special Report series. The volume will be supplied at a price to be announced later; one set of preprints will be sent to all those whose order for a bound volume is received by the Secretary before the meeting. Preprints of papers will NOT be available separately.

Special meeting in USA October – November 1961

The Council hope shortly to be able to announce full details of the programme for the Institute's Special Meeting in the USA and Canada next year. As previously announced, the Council has accepted with pleasure an invitation from the Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers (AIME) to visit the USA in the autumn of 1961.

The provisional programme provides for an official welcome in New York City on **Thursday, 19 October 1961,** but it is anticipated that some members will arrive a day or two earlier.

Members will visit important works within convenient distance from New York. They will then travel to Niagara Falls, and visit laboratories and works in the district in Canada as well as in the USA. Members have been invited by the Board of Directors of the American Society for Metals (ASM) to attend the National Metal Congress in Detroit and to be guests at the annual banquet of the ASM on 26 October.

The itinerary will include Cleveland, Pittsburgh, Bethlehem, and Washington. Opportunities will be provided for members to visit iron- and steelworks and laboratories in various centres. A special programme for ladies is being prepared.

At the conclusion of the tour, arrangements can be made for members to spend a few days sightseeing, visiting Philadelphia, and Princeton or Colonial Williamsburg and Charlottesville. Extensions will also be arranged to enable members to visit works in the Chicago district.

Members will be able to travel to and from the USA by normal sea and air routes, but in addition aircraft will probably be chartered for Atlantic crossings at substantially reduced cost. Provisional dates for these crossings are: to New York, **Wednesday, 18 October;** from New York, **Wednesday, 8 November 1961.**

Full information and quotations of costs will be circulated in the spring of 1961.

Correction

In the 'News' section of the *Journal* for July, Professor W. S. Owen, of the Department of Metallurgy, University of Liverpool, was wrongly shown as being with Sheffield University.

NEWS OF MEMBERS

Mr **C. R. Wheeler,** C.B.E. (*Past-President*) is to relinquish the chairmanship of Guest Keen Iron and Steel Company Ltd on 31 December next. He has been appointed an additional vice-chairman of Associated Electrical Industries Ltd and will take up executive duties with AEI in 1962. Mr Wheeler will also relinquish his directorships and executive appointments with the GKN group at the end of the year, although he will remain a director of Guest Keen and Nettlefolds Ltd.

Mr **W. H. Adams** has been appointed general manager of Head Wrightson Iron and Steel Works Engineering Ltd.

Dr **Howard Biers,** senior consultant to the Union Carbide International Company, is now at the company's head office in New York.

Mr **H. Darnell** has been appointed general works manager of the Workington Iron and Steel Company, and a director of the company.

Mr **J. O. Edström** is now director of Sandvikens Jernverks AB,

Mr **Ambrose Firth** has retired from the chairmanship of Brightside Engineering Holdings Ltd, for reasons of health.

Mr **T. C. Firth** has been elected chairman of Brightside Engineering Holdings Ltd.

Dr **S. Garber** is now with the College of Engineering, University of Saskatchewan, Canada.

Mr **A. E. Gilbert** has been appointed managing director of the Edmonton Steel Strip Co. Ltd. He has also been elected chairman of council of the Institute of Sheet Metal Engineering.

Lord **Dudley Gordon** is to retire as chairman and from the board of J. and E. Hall Ltd, on 30 September.

Mr **H. M. Henderson** has been appointed director and general manager of the Workington Iron and Steel Company.

Mr **Andrew Jollie** has resigned from Steel, Peech and Tozer.

Mr **W. W. Kee** has been elected chairman of the Association of Light Alloy Refiners and Smelters Ltd.

Mr **H. F. Maton** has been appointed managing director of the Harlow Metal Co. Ltd.

Mr **D. Manterfield** has retired from the post of technical consultant to Steel, Peech and Tozer.

The title of Bergrat h.c. has been conferred by the President of the Austrian Republic on Dr. mont. Ing. **Bernhard Matuschka.**

Mr **C. D. Muntz** has been appointed a joint managing director of the Woodall-Duckham Construction Company Ltd.

Dr **T. E. Peacock** has joined the chemistry department of University College, London.

Mr **F. A. Platts** has been appointed London office manager of The United Steel Companies Ltd.

Mr **R. O. Richards** has been appointed director-in-charge of the design and development department of the Woodall-Duckham Construction Company Ltd.

Mr **G. E. Robinson** has been appointed general sales manager of The Brightside Foundry and Engineering Co. Ltd.

Mr **T. Sanderson** has been appointed deputy general manager of the Workington Iron and Steel Company.

Mr **K. A. Smith** has been appointed production director of Firth Cleveland Steel Strip Ltd.

Mr **H. A. Snow** has been appointed industrial liaison officer with the Coil Spring Federation Research Organization.

Mr **H. Southern** has been elected the first president of the Society of Furnace Builders.

Mr **H. F. Spencer** has resigned from the board of R.F.D. Ltd.

Mr **G. Thickett** is now works manager (heavy departments) of Samuel Fox and Co. Ltd.

Dr **R. F. Weil** has joined Metal Containers Ltd.

Mr **H. A. A. While** has been appointed general export manager of The United Steel Companies Ltd.

Obituary

A. V. Francis (*elected* 1955) of Wolverhampton in July.

Eugene Clifford Grace, ENG.D. (*elected* 1914, *Bessemer medallist* 1942) of Bethlehem, Pa., USA, in July.

P. M. Gunst (*elected* 1959) of Amsterdam, on 30 June.

Roy M. Hyslop (*elected* 1933) of London, in February.

G. Murray, M.SC. (*elected* 1936) of Oxford, on 16 June.

Stanley Robson, M.SC., D.I.C., M.I.M.M., F.R.I.C., (*elected* 1937), of London, on 26 March.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Retirement of Dr J. C. Hudson

Dr J. C. Hudson, D.SC., D.I.C., A.R.C.S., F.I.M., who has been in charge of the British Iron and Steel Research Association's research on corrosion since the formation of the Association in 1945, retired at the end of August. He will continue to have very close ties with the Association, since he has agreed to act as consultant to both the Corrosion Advice Bureau and the Chemistry Department.

Dr Hudson was educated at Brighton College and at the Royal College of Science to which he obtained a Royal Scholarship in 1920. After taking his degree and associateship, he undertook research in physical chemistry. His first appointment was with the Atmospheric Corrosion Sub-Committee of the British Non-Ferrous Metals Research Association, where he worked in close association with Dr W. H. J. Vernon, O.B.E. As a result of this work he was awarded the D.SC. degree of the University of London. Then, in 1929 he became investigator to the newly formed Corrosion Committee of The Iron and Steel Institute, since taken over by BISRA, and has continued in this post until now. He has also served as Head of BISRA's Corrosion Advice Bureau since its formation in 1954.

Dr Hudson has recently received two major awards in recognition of his outstanding contributions to corrosion engineering and ferrous metallurgy: the Frank Newman Speller Award for 1959 from the American National Association of Corrosion Engineers, and the Sir Robert Hadfield Bronze Medal for 1960 from The Iron and Steel Institute. He has published well over 100 original papers, including two books, 'The corrosion of iron and steel' and 'Protective painting of structural steel', the latter in collaboration with Mr F. Fancutt, M.B.E.

Mr J. F. Stanners, B.SC., succeeds Dr Hudson. He was educated at the Royal Grammar School, High Wycombe, and at Birkbeck College, London, where he gained the degree of B.SC.(Special) in Chemistry in 1944. From 1940 until 1945 Mr Stanners was engaged as an analytical and research chemist in work on the measurement of radioactivity at the research laboratories of Thorium Ltd (now the Radiochemical Centre) at Amersham. He then joined the staff of the Corrosion Committee of The Iron and Steel Institute, transferring to BISRA with Dr Hudson. Mr Stanners has acted for the last eight years as deputy to Dr Hudson and represents the Association on the Technical Panel of the Joint Committee for the Co-ordination of the Cathodic Protection of Buried Structures. He recently presented papers at two corrosion conferences in America.

Mr E. E. White, F.R.I.C., A.M.I.M.M. F.Z.S., M.I.INF.SC., has been appointed Head of the BISRA Corrosion Advice Bureau.

F. E. L. Copley

Dr A. A. Smith

Dr-Ing. R. Ebert

Prof. Dr-Ing. K.-F. Lüdemann

CONTRIBUTORS TO THE JOURNAL

F. E. L. Copley, A.MET., A.I.M.—Research staff, BISRA laboratories, Sheffield (*Contributor to the August issue*).
Mr Copley was born in Sheffield and educated at High Storrs Grammar School. His initial metallurgical training was with Brown–Bayley Steels Ltd, then in 1946 he joined the Parachute Regiment. He returned to Brown Bayley and later received his associateship in metallurgy at Sheffield University. In 1956 he joined BISRA and has worked chiefly on brittle failure of mild steel.

Dr-Ing. G. von Struve

G. von Struve, DR-ING.—Research Department, Lurgi Gesellschaft für Chemie und Hüttenwesen, Frankfurt/Main, Germany.
Georg von Struve was born in 1930 in Saxony, and attended college at the 'Bergakademie' in Freiberg, Saxony, where he passed his diploma examination as a metallurgist in 1954. For a year he worked as assistant to the blast-furnace controller at the Halmstadt ironworks, then he returned to the Bergakademie in Freiberg, first as assistant and later as chief assistant to the Institute of Ferrous Metallurgy. After graduating in 1956 Dr von Struve lectured on the metallurgy of pig iron and the construction of blast-furnaces. He joined the Lurgi Gesellschaft in 1950.

A. A. Smith, M.A., PH.D.—Lecturer in metallurgy, King's College, London.
Dr Smith was born in Birmingham in 1929 and educated at West House School, Birmingham, and Wycliffe College, Gloucestershire. He completed his national service with the RAMC as a radiographer before going up to Trinity College, Cambridge to read natural science in 1949. Dr Smith intended originally to specialize in chemistry, but later he took the Part II Tripos in metallurgy. From 1953 to 1956 he carried out research on mass transfer at Cambridge for which he was awarded his Ph.D. degree. For a short time he worked at Imperial College London on solubilities in liquid bismuth before taking up his present post in 1957. Dr Smith is married, and his chief interests outside metallurgy are music, antiques, and hill-walking.

R. Ebert, DIPL.ING.—Research manager, Eisenhütten-Institut, Bergakademie Freiberg, Germany.
Rolf Ebert was born in 1932 in Neustadt/Thüringen, Germany. From 1952 to 1957 he studied metallurgy at the Bergakademie, Freiberg, Saxony, where he gained his DIPL.ING. Since September 1959 he has been manager of a working group on burden preparation and pig-iron production at the Eisenhütten-Institut of the Bergakademie Freiberg.

K.-F. Lüdemann, PROF. DR.-ING.—Professor of metallurgy and director of the Eisenhütten-Institut, Bergakademie Freiberg, Germany.
Karl-Friedrich Lüdemann was born in 1912 at Lennep in Rhineland. From 1931 to 1936 he studied metallurgy at the Bergakademie Freiberg, then he worked in the iron and steel industry until 1950. He then joined the Eisen-Forschunginstitut at Hennigsdorf, near Berlin, as manager of the metallurgical department and assistant to Professor Dr-Ing. Eduard Maurer. In 1953 Dr Lüdemann was appointed reader at the Humboldt University in Berlin. He has held his present post at the Bergakademie Freiberg since 1956.

INSTITUT DE RECHERCHES DE LA SIDERURGIE

Monsieur Husson retires

Monsieur Gabriel L. Husson, head of the steelmaking department of IRSID since its establishment in 1947, has retired. He remains associated with IRSID in a consulting capacity.

Monsieur Bernard Trentini, M.SC, formerly Monsieur Husson's assistant, has been appointed his successor.

ELECTRIC STEELMAKERS' GUILD

Officers 1960–61

President: W. H. Everard
Vice-President: W. Keen
Secretary-Treasurer: R. S. Howes
Executive Committee: A. Mettam, S. E. Jarvis, E. Meggitt, J. W. Mitchley, E. Davies.
Honorary Auditors: Senior: F. C. Sunley, Junior: A. S. Auty

THE INSTITUTE OF PHYSICS

Final report

The report of The Institute of Physics for 1959 which was adopted at the final meeting of the Institute before its amalgamation with The Physical Society shows that at the end of the year the total membership in all grades was 6754, an increase of 445 on the previous year. The number of applications for membership continued to increase and reached a new record of 933 during the year. The report records that at a General Meeting on 1 December 1959 the scheme for amalgamation was unanimously adopted and the Board was requested to take the necessary action to implement the proposals.

The number of candidates entering for the Graduateship examination of the Institute continued to increase at a very satisfactory rate; 119 candidates took papers compared with 80 the previous year. Only 39 candidates were successful and of these 13 held university degrees and 26 the Higher National Certificate in applied physics.

The report again records the expanding activities in many fields including education, scientific publications, conferences and symposia, and meetings arranged by the branches and specialist subject groups. During the year the Institute's benevolent fund gave assistance in ten cases where members of the Institute or their dependents had met with misfortune.

EDUCATION

RoSPA training course

In the autumn the Royal Society for the Prevention of Accidents (RoSPA) plans to hold a one-week residential safety training

course for safety officers concerned with accident prevention in the building and civil engineering industries. As well as lectures on safety organization, causes of accidents, scaffolding, lifting appliances, and demolitions, there will be practical lectures on the features of building work which lend themselves to demonstration.

The course will be held at the Civil Defence Staff College at Sunningdale, Berks, from 13 to 18 November 1960. The inclusive fee for the course will be £25.

All enquiries should be made to the Training Officer, RoSPA, Industrial Safety Division, Artillery Mansions, 75 Victoria Street, London SW1.

City and Guilds of London Institute session 1960–61

New and revised schemes for the session 1960–61 have recently been issued by the City and Guilds of London Institute. Also available is a list of subjects in which examinations will be offered in 1961. Both may be obtained from 76 Portland Place, London W1.

FORTHCOMING CONFERENCES AND EXHIBITIONS

Seventh Commonwealth Mining and Metallurgical Congress

As announced earlier, this congress is to be held in southern Africa next year. The first part will be held in the Union of South Africa from 10 April to 6 May, and the second part will be held in Northern Rhodesia from 7 to 13 May. The sponsors say that the prime aim of the congress is educational, and they hope it will be a 'forum for the interchange of progressive views and new ideas on mining and other projects, proven and experimental, in their many and varied aspects; a meeting place for people of similar basic interests and with many common problems'.

The congress is open to members of the constituent organizations of the Commonwealth Council of Mining and Metallurgical Institutions, which in the UK include The Iron and Steel Institute, The Institute of Metals, The Institution of Metallurgists, the Institution of Mining and Metallurgy, the Institution of Mining Engineers, and the South Wales Institute of Engineers.

The second brochure of the congress, recently issued, gives further details, and is available from the Congress Manager, PO Box 809, Johannesburg, South Africa. A few applications forms for the congress are available from the Assistant Secretary, The Iron and Steel Institute.

Optical instruments and techniques

A conference on 'Optical instruments and techniques' is being arranged by the British National Committee for Physics, under the auspices of the International Commission for Optics. It will be held in London from 10 to 14 July 1961. Enquiries should be addressed to Mr K. J. Habell, National Physical Laboratory, Teddington, Middlesex.

Pipes, pipelines, pumps, and valves

The second international Pipes, Pipelines, Pumps, and Valves Exhibition and Convention will be held at Earl's Court, London, from 9 to 13 April 1962.

NEWS OF SCIENCE AND INDUSTRY

Guest Keen reorganization

The Guest Keen Iron and Steel Company has announced that from 2 January 1961 the separate steel companies within the group will be amalgamated under the name of GKN Steel Company. Mr K. S. Peacock will be chairman of the new company, which will acquire the undertakings and assets of the Brymbo Steel Works, Guest Keen Iron and Steel Company, John Lysaght's Scunthorpe Works, and the Scunthorpe Rod Mill (at present a subsidiary of John Lysaght's Scunthorpe Works). (*See also* 'News of Members', p.91.)

BSCRA Annual report, 1960

The seventh annual report of the British Steel Castings Research Association, recently issued, tells of a year of varied activity and outlines some interesting research projects being carried out by the Association. Oxygen injection has received further attention, and the Association's control charts, which define for various bath temperatures the amount of oxygen to be injected to produce a predetermined reduction in the carbon content, have had their validity substantiated by further foundry trials. Phosphorus removal during oxygen injection has also been studied, and there have been both pilot-plant and works trials on desulphurization of liquid steel by the 'reversed slag' process. Research into pinhole porosity will be facilitated when gas analysis apparatus is available at the Association's research station later this year.

The report also outlines the Association's research programme on foundry processes and moulding materials, metallurgy, plant engineering, industrial health, and non-destructive testing. Copies of the report are available from 5 East Bank Road, Sheffield 2.

Staff changes at Workington

The Workington Iron and Steel Company branch of The United Steel Companies Ltd recently announced a number of new appointments. Mr G. Pott (*Member*), at present works engineer, has become chief engineer. Mr W. Hunter (*Member*), blast-furnace manager, is appointed ironworks superintendent, with responsibility for ore preparation and blast-furnaces. Mr F. Greenalgh succeeds Mr Hunter as blast-furnace manager. Mr J. D. Young (*Member*), Bessemer manager, becomes assistant steelworks superintendent, and is succeeded as Bessemer manager by Mr G. Forster (*Member*), at present Bessemer shift manager.

Report of the National Chemical Laboratory 1959

Corrosion in the UK costs about £600 m. every year and even a slight improvement in preventive measures would represent considerable savings in materials and manpower, says the National Chemical Laboratory in its recent annual report. The report outlines ways in which the NCL is tackling this urgent national problem. Work on bacterial corrosion is being extended by research on factors affecting the aggressiveness of soils; a series of tests are being made on the corrosion of mild-steel ship plate in estuary water; an investigation of the corrosion of mild-steel tubes in Scottish marine boilers has been completed for the British Shipbuilding Research Association, and a pilot survey is being carried out on corrosion by polluted waters. The report, price 4*s* 6*d*, is available from HMSO.

Coking in colour

A new colour film, 'The Murton Coking Plant', jointly sponsored by the Woodall-Duckham Construction Company Ltd and the Durham Division of the National Coal Board, was recently previewed in London. The film divides into two parts, at first outlining the traditional coke-making processes which are shown to be slow, and wasteful of labour and materials. It then details the new plant with its high degree of centralized automatic control and instrumentation. Much of the film will be of interest to a non-specialized audience, although the elaboration of detail of the instrumentation and control is for more specialized viewing. The film is available on loan from the sponsors.

Fully automatic heat-treatment line

What is said to be the first completely automatic heat-treatment line in the UK for carburizing, annealing, hardening, and tempering has been installed by AEI-Birlec Ltd at British Timken's Daventry East works. The equipment has been built primarily for the heat-treatment of tapered roller bearing cups and cones for vacuum-brake rolling stock required under the railway modernization scheme.

The processes, which are in automatic sequence, are performed in four Birlec furnaces. The furnaces are linked by mechanical and hydraulic handling devices designed by British Timken to meet their specific requirements. The installation performs the following processes: gas carburizing, sub-critical annealing, hardening, and tempering.

New plant and equipment

West Instrument Ltd of Brighton have begun production of a multi-point scanning unit for nuclear energy, plastics, and general engineering applications. The instrument is designed for continuous monitoring of temperatures at up to 12 positions in succession, and sensing is effective above 200°C.

The Bren Manufacturing Company, Gateshead, a branch of the R. W. Crabtree organization, recently produced a new sheet-metal grader to the specifications of an American customer, the F. J. Littell Machine Company. The machine is to be used for trials in a complete coil-fed sheeting line for the US canning industry. It is fully automatic, the only manual handling being the removal of loaded pallets. It can handle sheets of from 18 in. × 24 in. to 44 in. × 38 in. at speeds up to 175 sheets/min, and will carry pallet loads of up to 9000 lb.

Contract news

Air Products (Great Britain) Ltd in association with *The Butterley Company* announce that they have completed building an oxygen plant at Stewarts and Lloyds' steelworks, Corby. The plant has an output of 200 tons of oxygen per day and is the first tonnage oxygen plant to be owned and operated by a steel company in the UK. Gaseous oxygen is produced at a pressure of 350 lb/in² and at 99·5% purity. Gas storage under pressure is provided for 370 000 std. ft³ of oxygen.

A contract for the supply of prestressed concrete pressure pipes and spun iron pipes worth about £400 000 has been placed with *The Stanton Ironworks Company Ltd*, near Nottingham, by the Cardiff Corporation Waterworks.

Dewhurst and Partner Ltd have received instructions from Richard Thomas and Baldwins Ltd for the supply of individual motor starters and contactor pattern lighting control panels for the Spencer Works at Newport, Mon. The individual motor starters covering ranges up to 90 hp will be of the direct-on-line pattern of heavy duty steelworks construction built in accordance with a specification compiled by the RTB engineers, and will incorporate frame mounting type H contactors and type ET thermal overloads with inherent single phasing protection. The lighting control panels will include attached pattern contactors in various sizes up to 300 A and will be arranged for hand or automatic operation.

Changes of address and title

Allen West and Co. Ltd have transferred their Birmingham office to Pearl Assurance House, Temple Row, Birmingham 2, (Birmingham Central 7831).

Chaseside Engineering Company Ltd announce that all service matters relating to Lancashire and Yorkshire will in future be dealt with by their Service Division at Philips Road, Blackburn, Lancs. (Blackburn 4065).

High Duty Alloys Ltd have transferred their Birmingham office to 369 City Road, Edgbaston, Birmingham 16 (Bearwood 2344–5).

F'AG Bearing Company Ltd have recently opened branch offices at 36 Mansel Street, Swansea and 35 Call Lane, Leeds 1.

The Witton Moulded Insulation Works of The General Electric Company Ltd will in future be known as GEC Moulded Plastics Division.

CORRIGENDA

Authors' corrections to the paper '*The effects of hydrogen on the mechanical properties of low-alloy rotor forging steels*', by R. M. Goldhoff *et al.* (*JISI*, 1960, **195**, July, 296–302) received after the issue had gone to press are as follows:

p.300, right-hand column, line 11, the hydrogen range should be 0·5–1·5 ppm;

p.301, left-hand column, first sentence should

read 'The points originally set down will now be considered individually'.
On the same page, Fig.4 should include the following sub-caption: 'Figures in parenthesis indicate hydrogen contents of specimens after rupture times shown'.

DIARY

31 Aug.–7 Sept. BRITISH ASSOCIATION FOR ADVANCEMENT OF SCIENCE—Conference—Cardiff.

4–11 Sept. INTERNATIONAL AUTUMN TRADE FAIR—Vienna.

5–9 Sept. INSTITUTE OF METALS—Autumn meeting—Bath.

6–16 Sept. PRODUCTION ENGINEERING SHOW Chicago, USA.

9–10 Sept. CZECHOSLOVAK SCIENTIFIC-TECHNICAL ASSOCIATION—Symposium on fatigue strength—Prague.

10–17 Sept. INTERNATIONAL INSTRUMENTS EXHIBITION—Stockholm.

10–25 Sept. GERMAN INDUSTRIES EXHIBITION—Berlin.

11–20 Sept. GERMAN MACHINE TOOL FAIR—Hanover.

11–25 Sept. INTERNATIONAL TRADE FAIR—Brno, Czechoslovakia.

15–18 Sept. BRITISH WIRE ROD ROLLERS ASSOCIATION—Conference—Scarborough.

15–26 Sept. INTERNATIONAL INDUSTRIAL AND TECHNICAL EXHIBITION—Charleroi, Belgium.

17 Sept.–2 Oct. INTERNATIONAL TRADE FAIR—Mons, Belgium.

21–23 Sept. VEREIN DEUTSCHER INGENIEURE—Meeting on clean air—Wiesbaden, Germany.

21 Sept.–1 Oct. FACTORY EQUIPMENT EXHIBITION—Belle Vue, Manchester.

22–24 Sept. INSTITUTE OF PHYSICS *et al.*—Conference on *Theory and practice of ultrasonic inspection*—Queens Hotel, Cheltenham.

26–28 Sept. FEDERATION OF BRITISH INDUSTRIES—Conference—Palace Hotel, Buxton.

26–30 Sept. INTERNATIONAL INSTRUMENT-AUTOMATION EXHIBITION—Pittsburgh, Penn., USA.

26 Sept.–10 Oct. NATIONAL INDUSTRIAL SAFETY WEEK.

27–28 Sept. THE INSTITUTION OF MECHANICAL ENGINEERS—Symposium on automatic control—1 Birdcage Walk, London SW1.

27–29 Sept. DSIR—Conference on ergonomics—Connaught Rooms, London WC2.

1–6 Oct. INTERNATIONAL TRADE FAIR—Metz, France.

3 Oct. INSTITUTE OF BRITISH FOUNDRYMEN, SHEFFIELD BRANCH—Presidential address—Technical College, Pond Street, Sheffield, 7 pm.

3–7 Oct. CANADIAN NATIONAL MATERIAL HANDLING SHOW—Montreal, Canada.

4–7 Oct. SOCIETY FOR CLEAN AIR—Conference and exhibition—Harrogate.

5–8 Oct. BRITISH STEEL FOUNDERS ASSOCIATION—Conference—Folkestone.

7–16 Oct. ITALIAN MACHINE TOOL EXHIBITION—Milan, Italy.

12 Oct. MANCHESTER METALLURGICAL SOCIETY—Presidential address, by A. B. Ashton, M.SC., F.I.M.—Manchester Literary and Philosophical Society, George Street, Manchester, 6.30 pm.

12 Oct. BRITISH NUCLEAR ENERGY CONFERENCE—Lecture '*The inspection of nuclear power plant*' by H. N. Pemberton—Institution of Mechanical Engineers, 1 Birdcage Walk, London SW1, 6 pm.

15–23 Oct. INTERNATIONAL ELECTRONICS EXHIBITION—Lubljana, Yugoslavia.

17–21 Oct. SOCIÉTÉ FRANÇAISE DE MÉTALLURGIE—Journées Métallurgiques d'Automne—Maison de la Chimie, 28 bis, rue Ste Dominique, Paris 7e.

17–21 Oct. NATIONAL METAL EXHIBITION AND CONGRESS—Philadelphia, USA.

18–19 Oct. VEREIN DEUTSCHER INGENIEURE—International conference on cogwheels and gearwheel drives—Aachen, Germany.

18–20 Oct. NATIONAL SAFETY CONGRESS—Central Hall, Westminster, London SW1.

18–20 Oct. ENGINEERING INDUSTRIES ASSOCIATION—13th Engineering Display—Royal Horticultural Society New Hall, London SW1.

18–21 Oct. NATIONAL INDUSTRIAL SAFETY, HEALTH AND WELFARE EXHIBITION—Granby Hall, Leicester.

18–21 Oct. FIRST EFFLUENT AND WATER TREATMENT EXHIBITION AND CONVENTION—Seymour Hall, Seymour Street, London W1.

19–26 Oct. INTERNATIONAL CONGRESS AND EXHIBITION FOR INSTRUMENTATION AND AUTOMATION—Düsseldorf, Germany.

20–21 Oct. BRITISH IRON AND STEEL RESEARCH ASSOCIATION—28th Blastfurnace Conference—Scarborough.

25 Oct. BRITISH NUCLEAR ENERGY CONFERENCE—Lecture, '*Some observations on the design and construction of the* N.S. Savannah', by J. A. Dodd and S. MacDonald—Institute of Marine Engineers, 76 Mark Lane, London EC3.

26 Oct. MANCHESTER METALLURGICAL SOCIETY—Meeting and lecture by Prof. C. R. Tottle (The University, Manchester)—Manchester Literary and Philosophical Society, George Street, Manchester.

28–30 Oct. INSTITUTE OF MATERIALS HANDLING—Conference—Peebles.

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on page 486 of the August 1960 issue of the *Journal*:

Polarized light in metallography. Application to the determination of austenite grain size.
MUÑOZ DEL CORRAL, F., *Inst. Hierro Acero* (Madrid), 1954, Oct.–Dec., pp.464–467. (£2 5*s.* 0*d.*) [**1531**

The determination of rated capacity and production control in the grinding and descaming of billets on the basis of statistical analysis of performance.
WOLF, J., *Stahl u. Eisen*, 1959, July 23, pp.1075–1079. (£3) [**1548**

The progress of solidification of mild steel ingots in ingot moulds.
PARMENTIER, P., and SEVRIN, R., *Rev. Univ. Mines*, 1959, vol.15, November, pp.783–795. (£5 10*s.* 0*d.*) [**1579**

Recent developments in the Sturzelberger iron reduction process.
REINFELD, H., *Radex Rund.*, 1951, Sept., pp.178–187. (£5 15*s.* 0*d.*) [**1597**

Research on austenitic Cr–Mn–C–N stainless steels. (Summary, giving 12 tables, 41 figures, and 60 references, with translated captions.)
HSIAO, CHI-MEI, *Acta Metallurgica Sinica*, 1958, (2), pp.138–180. (£12) [**1624**

Study on the hot-ductility of steels. I—study on the hot torsion test.
MORISHIMA, T., *Tetsu to Hagane*, 1958, May, pp.552–559. (£8) [**1631**

Ten years of the coke quality agreement. Its significance for coke-oven and blast-furnace operation.
DICKENS, P., and RADMACHER, W., *Stahl u. Eisen*, 1960, Feb. 4, pp.129–136. (£5) [**1645**

The presentation of the results of work sequence and transport studies, and punched card evaluation of them.
GRÖGER, E., *Stahl u. Eisen*, 1960, Feb. 4, pp.165–169. (£4) [**1646**

The forging process on hammers and presses with particular reference to rising (die filling).
STÖTER, J., *Werkstattstechn.*, 1959, April, pp.223–230. (£4 15*s.* 0*d.*) [**1650**

The development of modified 12%Cr steels for forgings and castings in steam turbines.
KOUTSKY, J., *Hutnické Listy* (Prague), 1959, (12), pp.1105–1107. (£5) [**1671**

Intermetallic phases in two types of modified 12% chromium steels.
PLUHAR, J., *et al.*, *Hutnické Listy* (Prague), 1959, (12), pp.1108–1111. (£6) [**1672**

Measuring the roughness of cold-worked surfaces.
LUEG, W., and KRAUSE, U., *Stahl u. Eisen*, 1960, March 17, pp.325–337. (£8 5*s.* 0*d.*) [**1684**

Investigations on the economic operation of D.C. reversing rolling mill motors for heavy rolling mills.
GÜNDEL, M., *Metallurgie u. Giesserei Technik*, 1955, Feb., pp.50–60. (£8) [**1691**

The introduction of the magnetic amplifier for the high-speed control of rolling-mill drives.
WETZGER, J., *Stahl u. Eisen*, 1955, April 21, pp.478–485. (£6 15*s.* 0*d.*) [**1700**

Measuring principles of the magnatest Q-method and its testing possibilities in the steel manufacturing industry.
MICHALSKI, F., and KEMENESY, E., *Materialprüfung*, 1960, (2), pp.56–64. (£5) [**1704**

Prestressed screws in steel construction.
STEINHARDT, O., *VDI–Z*, 1959, July 1, pp.769–773. (£4) [**1720**

Modern ways of organization and technical means of increasing the efficiency of transshipment and transport installations for steel production.
SCHIMKE, E., *Stahl u. Eisen.*, 1960, Jan. 7, pp.12–19. (£6) [**1730**

A contribution to the plastic forming of internal threads.
KRETSCHMER, R., *Werkstattstechnik*, 1960, Jan. pp.12–14. (£2 15*s.* 0*d.*) [**1739**

Considerations of economy in the choice of the charge in open-hearth steel process.
SCHMIDT, G., *Stahl u. Eisen*, 1960, Mar. 3, pp.282–285. (£2 15*s.* 0*d.*) [**1742**

Operational investigations on the influence of the drawing temperature on the mechanical properties of drawn steel.
SCHWEIR, F., *Stahl u. Eisen*, 1959, Oct. 1, pp.1385–1391. (£4 15*s.* 0*d.*) [**1743**

Development of scrap iron quantities and prices in the European coal and steel Community.
GÜLDNER, W. A., *Stahl u. Eisen*, 1960, Jan. 21, pp.90–101. (£6 5*s.* 0*d.*) [**1745**

Mains network disturbances by arc furnaces.
WALTER, F., *Elektrotechnik*, 1948, August, pp.217–220. (£4) [**1746**

Pseudo-random numbers for the 'Strela' computer.
SOBOL', I. M., *Teoriya Veroyatnostei i ee Primeneniya*, 1958, (2), 205–211. (£4) [**1753**

The effect of pre-stressing in roll housings on roll accuracy.
PAWELSKI, O., *Stahl u. Eisen*, 1959, Sept. 3, pp.1295–1297. (£1 15*s.* 0*d.*) [**1773**

Increased economy by the use of precision castings.
NICKEL, E. G., *Industrie–Anzeiger*, 1960, *82*, May 10. (£4 10*s.* 0*d.*) [**1776**

Investigations on the activity of carbon in crystallized binary and ternary iron–carbon alloys.
SCHENCK, H., and KAISER, H., *Arch. Eisenhütt.*, 1960, April, pp.227–235. (£6 5*s.* 0*d.*) [**1779**

The influence of phosphorus and nitrogen on the hardness of low-carbon rimming steel after different pre-treatments.
DREVERMANN, A., *Stahl u. Eisen*, 1960, *80*, June 9, pp.801–809. (£5 15*s.* 0*d.*) [**1789**

A contribution to the investigation of pressure losses in pneumatic handling systems (with special reference to the influence of friction and weight of the material dealt with).
HAHNEMANN, H. W., *VDI–Z*, 1960, Feb. 1, pp.131–132. (£2) [**1790**

Direct observation of metal surfaces by reflection.
FERT, C., Electron microscopy: Proc. of Stockholm Conference, Sept., 1956, pp.8–12. (£2) [**1802**

On the character of the hysteresis of the demagnetizing factor.
HAJKO, V., and DANIEL-SZABO, J., *Czechosl. Journ. Phys.*, 1959, *8*, pp.685–688. (£2 5*s.* 0*d.*) [**1815**

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

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* Abstracts, the reference to which is marked with an asterisk, are or are to be available as translations.

ABSTRACTS

These abstracts are also available on index cards in advance of publication in the *Journal*, each abstract being classified under the Universal Decimal system (UDC). Details and subscription notes can be obtained from the Secretary, The Iron and Steel Institute.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Investigations on the tuyere and combustion zones in blast-furnaces and low-shaft furnaces R. Ebert and K-F. Lüdemann (*JISI*, 1960, **196**, Sept., 46–50) [This issue].

The acid smelting of foundry pig iron in low-shaft furnaces G. von Struve (*JISI*, 1960, **196**, Sept., 50–56) [This issue].

PRODUCTION OF STEEL

Durgapur Steelworks (*Iron Steel*, 1960, **33**, May 25 (special issue), 282–312) A fully illustrated account and progress report with plans.

Modernization of the Seattle plant N. L. Bock and K. Grant (*Iron Steel Engr.*, 1960, **37**, Feb., 77–87) Details are given of the construction and modernization programme recently completed at the Seattle plant of Bethlehem Steel Co.

Natural gas moves into steelmaking R. A. Labine (*Chem. Eng.*, 1959, **66**, Dec. 28, 50–53) A process flowsheet on the HyL process.

The economic effects of modernization in metallurgy S. Kawinski (*Prz. Techn.*, 1960, **81**, (7), 10–11) From data available the author has calculated a return of 5·95% for the old type of steelworks and a theoretical return of 20–40% for the modern installations. In general the output of modern works is 2·5 times greater than that of the older type. However, since the building of a new works is a lengthy process involving large outlays in capital, the author concludes that the building of new works should go side by side with the modernization of existing works.

Rotor in South Africa (*Met. Bull.*, 1960, April 26, supp. I, IV) Illustrations of the ISCOR plant.

Basic principles of combustion—model research A. A. Putnam and E. W. Ungar (*Trans. ASME J. Eng. Power*, 1959, **81A**, Oct., 383–388) The principles of the design and use of cold-flow models for the study of combustion equipment are set out. Various properties of burners are considered.

Waste heat recovery in metallurgical furnaces I. Stoicescu and C. Stamatescu (*Met. Constr. Masini*, 1959, (11), 940–949) The hot gases which leave the hearth between 1400 and 1750°C still have, after passing through the regenerators, a substantial amount of waste heat. Boilers suitably placed can recover this. The boilers are generally also connected to the furnace cooling water system for the purpose of heat recovery. Schemes are given for suitable installations.

Use of oxygen in the iron and steel industry M. Allard, Emery, Husson, Leroy, B. Trentini, and Blain (*Techn. Mod.*, 1959, **51**, Dec., 585–591) A review of the use of O_2 in blast-furnace, OH, and converter practice, and in the rolling mill.

Refinement of iron with oxygen and with powdered lime B. Trentini, P. Vayssiere, J Français, and M. Allard (*Fonderie*, 1960, March, 79–88) Refinement in the Bessemer and OH furnaces is described, using a suspension of lime in O_2. A method of partial refinement for removal of P and Si is then given, using a mixture of lime flux and iron oxide suspended in O_2. The theory of these methods is discussed (12 refs).—R.P.

Oxygen and steelmaking processes: a comparison of open hearth, LD, Kaldo and Rotor processes E. F. Kurzinski (*Iron Steel Engr.*, 1960, **37**, Feb., 65–75) Probable contributions of these processes to the anticipated 10% increase of world steelmaking capacity by 1965 are assessed.

Treat steel with magnesium to ease sulphur problem W. B. Brooks, A. C. Doumas, and B. W. Romefelt (*Iron Age*, 1960, **185**, Feb. 11, 148–149) Recent development work on the desulphurization of steel with Mg is reported. Work has been done in 500 lb basic electric arc furnaces. If up to 5 lb of Mg powder per t of steel, mixed with lime, is injected into steel, rapid S reduction ensues. Ductility and toughness properties are considerably improved, and large savings are possible. Further work is required on other properties and other steels, and on a larger scale.—D.L.C.P.

The mechanism of the desulphurization of iron by solid lime M. I. Gromov and L. M. Tsylev (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 25–28) This study has been carried out as a contribution to the elucidation of the mechanism and kinetics of the desulphurization of the metal. As a result, the conclusion is reached that desulphurization of the liquid iron by solid lime proceeds according to the scheme: $4\ CaO + 2\ S_2 = 4\ CaOS$. $4\ CaOS = 3\ CaS + CaSO_4$ between the two liquid phases metal and slag. Any difference in the reactions between liquid and solid phases will be apparent only in the kinetic and not in the static process.

Measurement and control in iron- and steelmaking practice H. Weineck (*VDIZ*, 1959, **101**, Oct. 21, 1415–1420) A review of the literature (85 refs).

Melting 19G and 14 KhG low-alloy steels without deoxidizing in the furnace with silicon I. M. Leikin, M. P. Sabiev, and V. A. Zhidkov (*Stal'*, 1960, (3), 216–219) In the process described saving of Si–Mn and deoxidants was achieved reducing Mn consumption by 8·8 and Si by 7%. Using Fe–Mn alone gave lower P and deoxidation in the ladle gave greater homogeneity and lower H_2 content; tapping temp. should be 10° higher. Impact strength was unimpaired and non-metallic inclusions were somewhat lower.

Basic oxygen steelmaking J. E. Tredennick (*Blast Furn. Steel.*, 1960, **48**, Jan., 51–57) The author reviews the establishment of the basic O_2 steelmaking process with particular reference to the furnaces at the Kaiser Steel Corporation plant at Fontana, California.—T.G.

The wheel mixer and granulometry of mixtures for (converter) bottoms M. Delayen (*Centre Doc. Sidér. Circ.*, 1959, **16**, (4), 841–845).

The design and construction of the new basic Bessemer plant at Port Talbot J. W. P. Jaffé (*Structural Eng.*, 1959, **37**, Oct., 282–291) The structural steelwork, foundations, crane guiders, and telpher system are described in detail.

Measuring and control plant at the experimental converter of Phoenix-Rheinrohr at Duisberg-Ruhrort K. Ewe (**Regelungstechnik*, 1959, **7**, July, 248–253) A description is given of the development work carried out with a 16-t converter by Phoenix Rheinrohr on the PL process, in which the metal is partially bottom-blown with air, then top-blown in the horizontal position using an oxygen lance, with a final bottom blow in the upright position using air or an oxygen-enriched blast. The measuring and control equipment used with the process is then described. This is basically an automatic O_2 and airflow measuring system, which is used as a mixture control system.

Oxygen process gets boost from refractory research R. E. Birch (*Steel*, 1960, **146**, Feb. 22, 76) The following linings have helped to make steelmaking in oxygen converters a success: permanent lining—burned magnesite material laid in periclase bonding mortar against the shell; working lining—tar bonded basic brick; intermediate lining—basic ramming mix. Better refractories will be available as the process develops.—D.L.C.P.

Conversion of phosphoric pig iron using solid finishing slag in the charge O. V. Travin, A. M. Skrebtsov, V. F. Surov, M. T. Bul'skii, P. N. Slepkanev, V. I. Alekseev, F. F. Sviridendo, V. A. Kostyuk, O. I. Val'ter, N. A. Kislov, B. A. Sharov, A. I. Osipov, and L. A. Shvartsman (*Stal'*, 1960, (4), 302–303) Addition of slag to the charge to accelerate removal was tried in 350-t furnaces using ^{89}Sr to follow slag formation. Removal takes place in the early stages and addition of 12 t of slag is recommended.

Preliminary experiment on the elimination of chromium from molten pig iron by Bessemerizing T. Soma, M. Tate, and K. Kanamori (*Tetsu to Hagane*, 1959, **45**, Nov., 1241–1247) Experiments in a 150 kW high frequency furnace, using blown air or oxygen enriched air,

show that Cr elimination is favoured by low temp. and low Si content. O_2-enriched air is preferable. The initial Mn content should be <0·8%.—K.E.J.

Processing of open-hearth and basic off-grade pig-iron in the oxygen blown converter Z. Zdeněk, K. Kučera, and O. Engel (*Hutn. Listy*, 1959, **14**, (12), 1073–1077) The steel, produced in a 5-t experimental oxygen-blown converter, with OH and basic Bessemer irons containing 0·5–1·1%P, was found to be equivalent in mechanical properties, and in fact slightly superior, to similar steels made in the OH furnace. The technology is described, and yield points, strengths, and ductilities of steels made by both processes are given and discussed. The steels made contained approximately 0·08%C, 0·35%Mn, 0·02%Si, 0·02%P, and 0·02%S.—P.F.

High pig-ratio operation in an LD converter S. Maehara, T. Kohtani, and K. Tagiri (*Tetsu to Hagane*, 1959, **45**, Sept., 959–961) The effects of adding mill scale and flux on operating parameters and production are quantitatively examined.

Distribution of arsenic and antimony between the liquid iron and the iron slag M. L. Sazonov and L. A. Shvartsman (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 28–31) Experimental findings confirm that As and Sb dissolved in the iron, form, during oxidation, unstable compounds, trioxides, with weak amphoteric properties and it follows that the distribution coefficients of these elements between metal and slag are very small. It also follows that one may not expect any appreciable elimination of these harmful elements from the iron through their passing into an oxidizing slag with very large basicity, and even less into an acid slag.

Elimination of hydrogen from molten steel by fuel cut-off in open-hearth practice M. Imai, H. Okabe, T. Nakayama, H. Ooi, and A. Ejima (*Tetsu to Hagane*, 1959, **45**, Sept., 955–957) The variation of the constitution of the furnace atm. and of moisture in it is shown for up to 30 min after cutting off the fuel. Operating results show the tendency to reduction of H_2 in the molten steel for up to 20 min after.

Effect of calcium oxide on the distribution of tungsten between liquid iron and slag B. M. Mogutnov, N. N. Perevalov, and L. A. Shvartsman (*Izvest. AN Metallurg. i Topl.*, 1959, (1), 22–28) The heat of the oxidation reaction of W by the iron-containing slag and by the iron and calcium containing slag with from 5 to 34% by weight of CaO, has been determined. It amounts to 14 800 cal/g atom for the iron-containing slag and varies from 21 800 cal. for the 5%CaO to 41 000 cal. for 34%CaO content of slag. The calculations of the heat reaction of WO_3 with the iron–calcium slag shows that this oxide has a definite acid character and therefore, with increased basicity of the OH slags, the oxidation of tungsten from the liquid iron must increase.

Distribution of niobium between iron and iron-containing slag M. L. Sazonov and L. A. Shvartsman (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 34–36) The experiments have determined the thermodynamical functions of the oxidation reaction of Nb dissolved in the liquid iron. The transition heat of 1 g atom of Nb from the iron into the iron-containing slag equals —70 500 cal. A comparison with findings from thermodynamical data justifies the conclusion that Nb is present in the iron-containing slag as the pentoxide.

Aerodynamics of the open-hearth furnace as a means of controlling the processes of combustion and heat exchange M. A. Glinkov (*JISI*, 1960, **196**, Sept., 1–14) [This issue].

The effects of flow distribution on air preheat in the open-hearth furnace J. Chapman and W. Montgomery (*JISI*, 1960, **196**, Sept., 15–28) [This issue].

The influence of deoxidation of steel and its treatment with sodium silicate on the content of non-metallic inclusions I. P. Kovyryalov, S. I. Popel', G. F. Konovalov, and A. M. Polzunov (*Stal'*, 1960, (4), 305–307) A rimming of OH steel deoxidized in the furnace with Fe–Mn and bottom poured for tinplate had 0·03–0·07% inclusions. Adding 2 kg/t Fe–Si in the furnace decreased this by one-half to one-third owing to increased slagging of lower-melting components. Addition of Na_2SiO_3 (24·1% Na_2O) to the tapping stream or to the mould (4 kg/t) further reduced inclusions and reduced the sticking tendency of the rolled sheets.

Studies in high-temperature gas cleaning: the reduction of fumes from open-hearth furnaces W. Strauss and M. W. Thring (*JISI*, 1960, **196**, Sept., 62–65) [This issue].

On the application of the Venturi scrubber to an open-hearth furnace T. Imaida (*Tetsu to Hagane*, 1959, **45**, Sept., 939–942) The layout of a cleaning plant is described, and detailed operating results are tabulated.—K.E.J.

Setup pre-refines molten iron for use in electric furnace Brymbo Steel Works Ltd (*Iron Age*, 1960, **185**, Jan. 21, 81–83) The steelmaking process used at Brymbo Steel Works is described; pig iron and scrap are treated in a pre-refiner before being added to molten scrap in an electric arc furnace. The blast-furnace iron is too low in phosphorus to be pre-treated in a Bessemer converter. The pre-refiner is a deep circular shaped tilting vessel of 20 t capacity with oxygen lances and tangential oil burners. Powdered lime is fed into the bath with the oxygen. Power consumption for the process is 395 kwh/t, and O_2 consumption is 460 ft^3/t on the electric furnace and 900 ft^3/t overall. The method is comparatively cheap and fast and has advantages over ordinary duplex treatment.—D.L.C.P.

Electric melting of steel H. W. A. Waring (*Financial Times*, 1959, Jan. 27, 6). **Modern arc furnace design** A. G. Robiette (7). **Mains frequency steel making** W. Kelsey (7).

The control of large electric arc furnaces in relation to their effect on the supply system J. W. S. Payne (*Instr. Practice*, 1959, **13**, Aug., 808–814) Furnace load and flicker is considered with a review of types of furnaces and systems of regulation. Field tests are reported and general conclusions are set out.

On improving the productivity of electric arc furnaces Z. Bužek, V. Schindlerová, and M. Benda (*Hutn. Listy*, 1959, **14**, (11), 931–936) An analysis of the performance of electric steelmaking furnaces used in Czechoslovakia and abroad is made. It is concluded that productivity could best be increased by shortening the melting period. This could be achieved by full utilization of transformer capacities, and the determination of the optimum current regime during steel production, at least for the steels most frequently produced in any given plant.—P.F.

Basic electric furnace practice for low alloy steels U. K. Bhattacharya and B. K. Gupta (*Inst. Brit. Found. Australian Branch*, 1959, **10**, 49–55) The uses and composition range of low-alloy steels are mentioned, and the steelmaking practice is discussed. The need for precise control of pouring and melting, and the special techniques required in the addition of some alloying elements are emphasized.

Investigation of the mechanisms in a DSP 80 arc steel melting furnace A. I. Sapko (*Stal'*, 1960, (4), 314–318) The furnace is described and its equipment and operation are reviewed. Only the mechanism for raising and lowering the electrodes is not altogether satisfactory.

Results obtained from 165-ton electric furnace G. Stassin (*Blast Furn. Steel Plant*, 1959, **47**, Aug., 862–864; 885) The layout, operation, and results obtained from the 165-t, 24 ft dia. Lectromelt electric furnace installed at the plant of Fabrique de Fer, Monceau-sur-Sambre, Belgium in 1956 is described. The furnace is top-charged in three to five stages and is equipped with a magnetic stirrer. Melting-down is assisted with O_2.

On the behaviour of hydrogen in molten steel. IV. On the behaviour of hydrogen in molten steel through the oxidizing period of basic electric arc furnace process. Part 2 S. Nagata, K. Takinami, T. Sugiyama, and S. Sato (*Tetsu to Hagane*, 1959, **45**, Sept., 966–967) The effects of adding CaO, limestone, and fluorspar after O_2 blowing and before slag-off, and of the kind of tool used for slag-off, are described.—K.E.J.

A new method for ventilating arc furnaces for steel melting A. Christea and V. Drobota (*Met. Constr. Masini*, 1959, (12), 1072–1076) Diagrams are presented of a scheme which has been tested in practice and for which considerable advantages are claimed, such as improvement of the working conditions and reduction of air temp. in the working space with increased air circulation. 50% savings in electrical power and simplicity of operation etc. are claimed.

Bibliography on the automatic control of electric arc furnaces (*Bibliography*, No.34, 1959, March, pp.3. *Supplement No.1*, July, pp.2).

Fundamental study of automatic control. On temperature measurement in an electric arc furnace A. Yamaki (*Tetsu to Hagane*, 1959, **45**, Sept., 964–968).—K.E.J.

Report of the 'ingot casting' sub-committee meeting of the Commission des Aciéries Thomas, 30 Jan. 1958 (*Centre Doc. Sidér. Circ.*, 1959, **16**, (4), 893–894). Brief abstracts are given of papers presented.

Increasing the life of ingot moulds N. F. Troshin and V. K. Sotnikov (*Stal'*, 1960, (4), 380–382) Reinforcement with welded steel bands of rolled sections and the use of iron with Cr and Ni and reduced Si content to reduce erosion show improvements. The use of Mg might improve the iron further.

Studies on capped steel ingots. I. Studies on the surface conditions of capped steel ingots T. Kato, J. Imai, Y. Hirose, and H. Kajioka (*Tetsu to Hagane*, 1959, **45**, Oct., 1139–1144) Using 2, 4, and 8 t bottle-top moulds, test ingots of capped steel (0·06–0·09%C, 0·36–0·50%Mn) made under various conditions were tested. The thickness of solid skin was controlled by Al addition in the ladle and pouring rate. The thickness of the rim zone depended on the hitting time. Cracks were reduced by reduction of Al in the ladle and improved design of mould (13 refs).—K.E.J.

Studies on capped steel ingots. II. Studies on the core zone in capped steel ingots T. Kato, J. Imai, Y. Hirose, and H. Kajioka (*Tetsu to Hagane*, 1959, **45**, Nov., 1261–1264) Segregation is less than in a rimmed ingot. The degree of max. segregation increased with Al added in the ladle, but was less pronounced than in a rimmed ingot.—K.E.J.

Ingots cast with controlled rimming in bottle-type moulds. Results obtained on production at Rehon G. de Rousiers (*Centre Doc. Sidér. Circ.*, 1959, **16**, (4), 895–897) Yield, degree of segregation, and surface quality of rolled products are compared with similar data from normal ingots.

Stirring trials on killed steel in ingot moulds R. Périé (*Centre Doc. Sidér. Circ.*, 1959, **16**, (4), 899–902) Gas is passed through the base of the bottom plate into the ingot during pouring and for 15 min after filling; then anti-piping compound is added to the ingot surface. Pipe and axial segregation are greatly reduced, and a uniform fine crystalline structure is obtained. The best results are obtained with CO, although the experiments described were with N_2 (which gave a small N_2 pickup).

Elongated ingots V. M. Tageev and V. A. Dudkin (*Stal'*, 1960, (4), 304–305) Two forms are described reducing feeder head volume.

Oxidation of rimming steel ingot during teeming and solidification Wang Ching-huei, Chang Shun-nan, Tung Ling-sung, Shih Chang-hsu (*Acta Met. Sinica*, 1959, **4**, Dec., 287–296) Rimming action was tested in O_2, N_2, water-gas, and air and also with control of admission of air to the ingot. Secondary oxidation appears to be the source of O_2. Regulation of the admission of air controls rimming action and reduces segregation (tables in English).

On the erosion of pins in the graphite stopper head Y. Isono, N. Tukiyama, and M. Amagai (*Tetsu to Hagane*, 1959, **45**, Sept., 950–952).

The use of thinning fluxes to reduce the contamination of rimming steel by slag R. Z. Elimelakh, V. A. Machkovskii, and E. S. Shlyakhovetskii (*Stal'*, 1960, (3), 219–220) Trials on the pouring of large ingots were carried out and ^{45}Ca was used as tracer. Mould filling rates of 0·16–0·22 m/min with addition of fluxes to the metal surface gave greatly reduced deep contamination.

Experience gained in the 'Kosciuszko' Steel Works in top-casting K. Levittoux, S. Bednarczyk, and T. Chrzaszcz (*Hutnik*, 1960, **27**, (2), 65–75) In view of the restricted space in

one of the departments of the 'Kosciuszko' works top-casting was introduced to speed up and increase production. The authors give the results of tests carried out, on a 60 t experimental casting unit, casting once per day for two months. As a result a method of operation for the future is given such that wear on the ingot-moulds is reduced, losses are reduced, and the optimum rate of casting is obtained.

A general review on continuous casting of steel Chao Hwei-tian and Chang Shun-nan (*Acta Met. Sinica*, 1959, **4,** Dec., 331–350) Continuous steel casting machines are illustrated and reviewed and suggestions for the improvement of ingot quality are made.

Study on the continuous casting of steel. II. On secondary cooling in continuous casting Y. Aketa and K. Ushijima (*Tetsu to Hagane*, 1959, **45,** Sept., 967–968).—K.E.J.

Russian defence of the horizontal method of continuous casting and other remarks (*Report of Odessa Conference*, 1955, pp.11) Reports of discussion by representatives of the Kharkov Refractories Institute, Central Research and Development Institute for Ferrous Metallurgy, Hammer and Sickle Plant, Ministry of Shipbuilding, Ministry of Defence, and the Academy of Sciences.

A device for stopping continuous cast ingots in hydraulically driven withdrawal stands in case of emergency V. A. Karpov, A. Kh. Chernyi, and M. P. Levin (*Stal'*, 1960, (4), 329) Rolls are fitted which jam the ingot if the hydraulic pressure falls.

The formation of internal defects in square billets during the continuous casting of steel V. S. Rutes, N. A. Nikolaev, and V. I. Akhtyrskii (*Stal'*, 1960, (3), 212–215) Tracer studies on the depth of pipe and its relation to casting rate, and studies of porosity and the effects of rolling and secondary cooling are reported. The causes of these defects were found and measures to minimize them were adopted.

A study of pipes in semi-killed steel ingots T. Kato, J. Imai, and H. Kajioka (*Tetsu to Hagane*, 1959, **45,** Sept., 970–972) Data are given for pipe volumes, the relation of degree of deoxidation and moulded dimensions to thickness of the bridge, and the relation between time of teeming and rate of descent of hot metal (from the head of a 4-t ingot) with various mould dimensions.—K.E.J.

Effect of the quality of the pit-refractory for bottom pouring on the nature of the non-metallic inclusions in steel. I. Steel-melting tests in crucibles made of pit-refractory S. Nagata, T. Sugiyama, and M. Onishi (*Tetsu to Hagane*, 1959, **45,** Sept., 948–950) Operating results and cleanliness data are given for high-alumina, chamotte, and silica crucibles, with melting in air or in argon.—K.E.J.

Study on the formation and precipitation of deoxidation products E. Plockinger and M. Wahlster (*Stahl Eisen*, 1960, **80,** May 12, 659–669) The authors review the theories of the formation and precipitation of non-metallic inclusions during the deoxidation of steel melts. The studies on the mechanism involved in the formation of oxide phases in liquid steel were carried out with the aim of finding deoxidation agents which produce as few inclusions as possible in the steel during the precipitation oxidation. The effect of unavoidable secondary inclusions formed from the residual oxygen content of the melt is considered, as these inclusions may play a role in segregation phenomena (25 refs).—T.G.

The influence of the original structure of steel on its tendency to hot cracks O. D. Moldavskii and A. P. Pronov (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 47–51) A relation has been discovered between the tendency of low-carbon steel to the formation of hot cracks and the content of Al, Si, and Cr respectively The strength of steel during solidification varies between 0·4 and 0·8 kg/mm² and depends to a considerable extent on its original structure.

A study using radioactive isotopes of the formation of a skin on ingots T. M. Saar, V. A. Nosov, and I. G. Bolokhovskikh (*Stal'*, 1960, (3), 208–211) Using ^{35}S and ^{110}Ag with bottom-poured 4·5 and 5 t ingots with lunkerite in the feeder head, skin formation in the lower part of the ingot was studied. Possible causes were the slow filling of the lower part by the first portions of metal cooled in the runners, also oxidation of the metal in the runners if teeming is interrupted. Sinking of crusts was not observed. Rapid teeming greatly reduced the chance of crust formation.

On the improvement of the quality of ball-bearing steel. I. The effect of smelting and casting practice on non-metallic inclusions in ball-bearing steel Sih Tson-yow, Hsing Chung-shu, Cheng Shu-tang, Lee Chin-yuan, Chang Wen-po, and Li Her-sheng (*Acta Met. Sin.*, 1959, **4,** Sept., 177–194) In the basic electric furnace, non-metallic inclusions are reduced by the following practice. SiMn is added on slagging off before reducing; crushed CaC_2 is added to produce carbide slag, which is kept for over 40 min; FeSi or CaC_2 is added. The slag should be white, and is tapped before the steel. Finally, over 0·5 kg of Al per ton of steel is added. Linings and runners should be tar-soaked, and ferro-alloys should be free from Ti (19 refs).

PRODUCTION OF FERRO-ALLOYS

India expands ferromanganese industry L. Nahai (*Eng. Min. J.*, 1959, **160,** Oct., 112–113, 210) Location, ownership, capacity, and production of ferromanganese producers are tabulated, and a map shows the position of plants, ore deposits, and hydroelectric projects.

The economics of the production of low-grade ferrosilicon in blast furnace and in electric ferro-alloy melting furnaces V. N. Zimin and V. A. Shtanskii (*Stal'*, 1960, (3), 269–273) The economics of blast-furnace or electric melting are referred to a geographical basis in the USSR. In southern and eastern regions, electrotechnical methods are preferred.

Perovskite concentrate used in the production of ferrotitanium I. S. Kumysh (*Stal'*, 1960, (3), 230–233) A process eliminating the use of ilmenite and producing alloys up to 45%Ti is described. A good system for removal and cleaning of fumes is essential.

Activities of manganese oxid in silicate melts K. P. Abraham *et al.* (*JISI*, 1960, **196,** Sept., 82–89) [This issue].

FOUNDRY PRACTICE

When specifying a cast iron W. H. Moore (*Prod. Eng.*, 1959, **30,** Oct. 12, 65–67) Microstructures are illustrated and correlated with properties.

The future of cast iron E. Longden (*Iron Steel*, 1960, **33,** May 25 (special issue), 267–271) An illustrated account giving examples of its use, with a discussion of possible trends.

Casting design—A modern art A. R. Moore (*Mod. Casting*, 1959, **36,** Nov., 46–48) The advantages and limitations of castings are discussed.—A.D.H.

Design of cast parts from the point of view of tooling U. Skarbinski (*Fonderie*, 1960, Feb., 43–52) The author shows how tooling costs can be reduced by greater accuracy of castings, by reducing the time and number of operations and by reducing cutting times.—R.P.

Solidification of metals G. W. Form and J. F. Wallace (*Mod. Castings*, 1960, **37,** April, 145–155) The structure of the liquid and solid states, and the homogeneous and heterogeneous nucleation of a pure metal, and the subsequent growth of such nuclei are discussed. The discussion is extended to the solidification of solid solutions and eutectics, and methods of grain size control are then considered (65 refs).

Instrumentation in the foundry laboratory R. C. Rooney (*BCIRA J.*, 1959, **7,** Oct., 832–838) A review of modern instruments and their usefulness.—T.G.

International classification of ferrous cast metals A. Wittmoser (*Mod. Castings*, 1959, **36,** Dec., 47–54) Suggested reclassification includes the grouping of nodular graphite iron–carbon alloys as graphitic cast steel, malleable cast iron, and cast iron. Alloy group classifications are suggested (29 refs).

Coke producers promote iron castings (*Foundry*, 1960, **88,** Jan., 123) To promote increased use of iron castings, the American Coke and Coal Chemicals Institute, Washington, on behalf of the entire American merchant coke industry, will sponsor nine promotional, educational, and engineering films developed and to be produced by the Gray Iron Founder's Society, Cleveland.—T.G.

Interfacial tension and surface finish of sand castings J. F. Kayser (*Found. Trade J.*, 1960, **108,** Feb. 11, 161–162) An account of wetting phenomena and interfacial tension as related to foundry processes.

The importance of the statistical evidence for saving metal in foundries and forges A. Saria (*Met. Constr. Masini*, 1959, (11), 973–982) Methods are considered for organizing statistical data to show at a glance all deficiencies, especially as regards forges and foundries.

Cutting scrap losses through process improvement K. M. Smith (*Foundry*, 1960, **88,** Jan., 72–77) Process improvement is discussed and a number of faults and their causes are described.—T.G.

Casting loss reduction program H. W. Dietert (*Mod. Castings*, 1960, **37,** April, 172–175) A survey of causes of loss with notes on prevention.

Increasing yield, reducing the weight of the undressed casting, importance of co-operative work F. Sedlacek (*Giessereitechn.*, 1959, **5,** Aug., 235–236) An account is given of the introduction of centrifugal casting and co-operative work in improving the process of manufacture of cast ribbed plates.

A new test for fluidity V. Kondic (*Foundry*, 1959, **87,** Dec., 79–83) A new type of test is developed using the ability of a metal to fill a mould as a function of the section size of the mould. Four branches protruding from a common bar are used, all 17 mm wide but of various thicknesses (0·965, 1·27, 1·57, and 1·88 mm). Results obtained with this fluidity test are described.—T.G.

Desulphurization of cast iron by electrolysing slag North-East Institute of Technology, The Institute of Foundry Practice (*Acta Met. Sin.*, 1959, **4,** Sept., 270–281) The method is effective and cheap, particularly with >1%S. The S reduction increases with slag basicity. There is an optimum c.d. for the cathode (molten iron) and the amount of desulphurization increases with voltage and time (13 refs).

Carbon dioxide in foundry technique O. Gerstmann (*Giessereitechn.*, 1959, **5,** Sept., 270–272) Topics discussed are—the importance of the CO_2 process and its correct application; and the properties of CO_2 and methods of supplying it.

Removal of gases from ferrous alloys in precision casting P. Destable (*Mét. Constr. Méc.*, 1960, **92,** March, 139–153) The state of various gases in solution in the liquid metal, and their affinity for the various alloying elements which may be introduced into the composition of castings are described, together with a statement of principles to be applied in the production of sound castings.—S.H.–S.

The effect of bubbling methane gas through liquid cast iron on its castability L. Sofroni, C. Cosneanu, and E. Nicoloiu (*Studii Cercet. Metalurg.*, 1959, (4), 525–535) Experimental findings in the experiments on bubbling CH_4 through liquid cast iron have shown that as a result, linear contraction is reduced as is also the tendency to pipe formation. This is explained by the structural changes taking place during the bubbling of the gas.

3 Big 'M's: Modernization, Mechanization, Maintenance V. E. Lich (*Mod. Castings*, 1960, **37,** April, 36–40) Six case histories are given from the Granite City Foundry of the General Steel Castings Corp., showing the way in which mechanization and methodical maintenance save money. Pneumatic handling of bentonite, silica flour, corn flour, and moulding sand, the use of lift trucks and power conveyors, and the upkeep of belts in the foundry are included in these case histories.

'We make castings for foundries' W. R. Colsmann (*Mod. Castings*, 1960, **37,** April, 30–33) The Sarcol Foundry and Pattern Corp. have developed the Shaw Process to such an extent that as-cast tolerances in consecutive duplication are ±0·0025 in./in. for sizes 1–8 in. and ±0·0020 in./in. for sizes 8–16 in. The process is described, and illustrations of the finish obtainable are given.

Foundry equipment—will it save or spend

your money? D. M. Buchanan (*Mod. Castings*, 1960, **37,** Feb., 30–34) The economics of the installation and maintenance of new equipment are discussed, and a detailed account of depreciation, and some accounting methods used to take it into consideration, is given. The economic time to replace equipment, in view of the depreciation accounting technique employed, is also considered.

Description of an automatic moulding shop floor I. Lamoureux (*Fond. Belge*, 1959, **29,** (11–12), 275–279, 292) A large production, continuous, fully-automatic German plant is described and illustrated.

The total engineering approach to furnishing cast parts R. H. Herrmann (*Foundry*, 1959, **87,** Dec., 60–64) A scheme is described which a foundry has developed to assist its customers. It offers complete design, laboratory, and engineering help, a choice of moulding methods, heat-treating, etc.—T. G.

Link-Belt's new malleable-iron foundry (*Found. Trade J.*, 1960, **108,** April 7, 419–425) An illustrated account of the installation and equipment.

Leyland Motor's new pre-treatment plant caters for grey-iron and malleable castings (*Found. Trade J.*, 1960, **108,** March 31, 394–395) An illustrated account, including the phosphating process.

Brass foundry enters steel castings production J. M. Kaderly (*Foundry*, 1960, **88,** Jan., 116–120) The new foundry equipment for casting carbon steel, Monel, and corrosion-resistant stainless steel is described and illustrated.

Conveyance in foundries. Means of suspending stock (*Fonderie*, 1960, March, 111–114) Methods of suspending stock are discussed in relation to shape and size of stock.—R. P.

Reorganization of a fettling shop E. Holland (*Found. Trade J.*, 1959, **107,** Oct., 377–381) When it became necessary to re-equip a fettling shop with modern machinery, the shop layout was studied, in order to improve work flow, means of access, and safety of work. The changes made are described, and the results achieved up to the present are reported.

A contribution to the history of the use of oxygen in the production of iron P. G. Barinov, M. R. Pershin, G. D. Kovalenko, and N. E. Gubenkov (*Lit. Proizv.*, 1959, (11), inside back cover) Oxygen was first used in 1932 for enriching the cupola blast at the Khar'kov Locomotive Works on the initiative of A. F. Bondarenko, and a rough description is given of the apparatus constructed by him for enriching the blast to the extent of 2–2·5%. Since 1949, oxygen-enriched blast has been used continuously at these works, giving a liquid iron temp. of 1400–1420° with substantial savings in coke.

High-temperature properties of spheroidal graphite cast iron A. Plesinger (*Fonderie*, 1960, Jan., 18–23) High-temp. properties of cast iron are studied with a view to its use in place of carbon steel at up to 450°C. Mechanical properties and oxidation resistance were found to be satisfactory for this purpose (37 refs).

Equipment for modifying cast iron with magnesium N. I. Klochnev, V. T. Lisitsyn, B. S. Mil'man, G. S. Strizhov, I. O. Tsypin, and B. Z. Chernyak (*Stal'*, 1960, (3), 287) A drum is illustrated which can be sealed and is capable of rotation. Mg is weighed into a pocket and on turning over the drum it is forced in under its own vapour pressure. Additional pressure can be applied if necessary.

Nodular cast iron. II. Manufacture F. Rodriguez–Yufera y Muñoz (*Inst. Hierro Acero*, 1959, **12,** (59), 28–45) The theory of inoculation and the manufacture of ductile iron are described in detail.

How do you know it is ductile? G. F. Thomas (*Mod. Castings*, 1960, **37,** April, 182–184) The quality control of nodular iron castings used by the author's Company is described. The graphite structure is regularly checked metallographically to detect vermicular graphite, and Y-block test castings are regularly poured. Properties of test bars are also checked regularly.

Malleable iron—hot and cold (*Mach. Design*, 1959, **31,** Dec. 10, 136–139) A condensed account from the Handbook of the Malleable Founders Society giving tables of mechanical properties.

Commercial establishment in Australia of pearlitic malleable iron, its manufacture, control, physical properties, and applications R. R. Gibson (*IAAE J.*, 1959, **19,** Dec., 198–205) An account of the properties of malleable iron with illustrations of parts made from it.

Ductile iron castings versus carbon steel forgings and weldments J. L. Salbaing (*Mod. Castings*, 1959, **36,** Sept., 61–69) Advantages of ductile iron castings compared with plain C steel forgings and weldments can include a more suitable structure, improved strength and lower costs. A number of case histories is given in which advantages were gained by changing to ductile iron.

Ductile iron in the machine tool industry R. M. Bronson (*Prod. Eng.*, 1959, **30,** Design Digest Issue, Mid-Sept., 65–67) Properties are given and typical parts are shown.

Eight ways to design economy into malleable iron castings H. L. Kee (*Prod. Eng.*, 1959, **30,** Nov. 9, 82–93) Brief notes on properties and methods.

Solidification of steel castings C. W. Briggs (*Mod. Castings*, 1960, **37,** April, 157–168) This article is a review of the literature. The general principles of metal solidification are related to the case of steel castings, and the factors affecting the solidification of steel castings are discussed. The object of the review is to provide a sound engineering understanding of the principles involved (25 refs).

Increasing efficiency in the steel foundry G. D. McNair (*Brit. Found.*, 1959, **52,** Nov., 494–502) Methods control, the use of exothermic feeding materials and examples of CO_2 castings are dealt with.

Mn–V–Mo age-hardening austenitic steel foundry characteristics N. C. Howells and E. A. Lange (*Mod. Castings*, 1959, **46,** Sept., 55–58) Melting, casting, and heat-treatment techniques are described for austenitic 14–18%Mn age-hardening steels, containing 0·25–0·7%V and 0·4–3·5%Mo, and which can be aged to 250–500 Brinell and remain non-magnetic.

Contributions to the technology of casting A. Sarian (*Met. Constr. Masini*, 1959, (12), 1021–1032) Experiences are described with the casting of various types of steel wheels. The results obtained are preliminary and a large amount of co-operation is requested in order to work out a reliable procedure for the operations.

Application of the oblique casting process in steel foundries and the uses arising therefrom W. Sachse (*Giessereitechn.*, 1959, **5,** Aug., 234–235) The technique of casting with the mould at an angle of 10–30° is briefly described with practical details. Applications are given, and the increased production and economy achieved are instanced.

Canadian non-ferrous and iron foundry sand practice survey A. E. Murton (*Mod. Castings*, 1960, **37,** April, 199–208) A survey was conducted by the AFS Canadian Sand Committee (8–Y) of foundry sand practice. The results indicate that Canadian foundries have been gradually changing from natural to synthetic sands, mostly because they feel that this will give better control of their sand properties. Details of the complete survey, covering different types of foundry are given.

Relation between compressive strength of synthetic foundry sand and its composition T. Nangoh (*Imono*, 1959, **31,** Aug., 732–741) The sand is regarded as a two-phase system, sand grains, and moist clay. The water-clay ratio is also considered.

Pattern and core box equipment for blowing foundry sands Z. Madacey (*Mod. Castings*, 1960, **37,** April, 193–198) The factors upon which the venting and design of pattern and core boxes depend are discussed.

Hot properties of moulding sand A. G. Johnston (*Mod. Castings*, 1960, **37,** Feb., 38–40) The importance of an appropriate hot strength, minimum expansion, and high hot deformation is discussed, and ways of adjusting these properties by means of bentonite, varying sand size, clay, and water content are considered.

Crust separation test for investigating sand expansion defects P. W. Goad (*Inst. Brit. Found. Australian Branch*, 1959, **10,** 77–78) The development of this new technique for studying expansion defects is described. It has been shown that contrary to previous opinion, expansion failures can occur even when deformation is far greater than expansion. The vital part played by the moisture content of the sand in determining the nature of the expansion failure, and the susceptibility of the moulding, is emphasized.

Apparent thermal conductivity of moulding sands at elevated temperatures D. H. Whitmore and Q. F. Ingerson (*Mod. Castings*, 1960, **37,** Feb., 49–57) The apparent thermal conductivity of a synthetic silica moulding sand was determined in the range 900–1800°F. It was found that below 1100°F the conductivity decreased with increasing temp. A minimum occurred between 1100 and 1500°F, and the conductivity rose sharply above this temp. The results were interpreted in terms of a model which considers heat transfer both by pure thermal conduction and by radiation across the pores and grains of the sand, and it was shown that the predictions based on this model were very close to the experimental results.

Development of a CO_2-process moulding sand H. T. Bidwell (*Found. Trade J.*, 1959, **107,** Oct., 383–385) In order to overcome the low green strength associated with CO_2-process moulding sands, a method of avoiding the detrimental effects of clay and moisture on these sands was developed. When a readily ionized sodium salt was added, it was possible to use sand of a higher green strength, and this has been done very successfully in production.

Exothermic compounds in the production of heavy castings I. Stadler (*Brit. Steel*, 1959, **25,** Oct., 348–352) The use of Feedex exothermic addition, produced by Foundry Services Ltd. in the manufacture of turbine casings by Litostroj, Ljublijana, Yugoslavia, is described.

Gating of ductile iron castings R. W. White (*Foundry*, 1960, **88,** Feb., 101–107) The characteristics of ductile iron are explained, and the technique of gating to produce quality castings in this metal is described. The first step in designing a gating system is to determine the proper points of metal entry. The next step is to determine the speed with which the mould should be poured (a graph illustrates this) and then calculate the required ingate area which depends on the pouring weight and pouring time. Thirdly, the sprue and runner must be fixed, for which area ratios are given. Two gate systems can be used —(*a*) the positive pressure system in which the choke occurs at the junction of the runner bar and ingate, (*b*) tranquil flow or reversed choke system in which the choke occurs immediately after the sprue.—D. L. C. P.

Blind feederheads and Washburn cores M. Notte (**Fond. Ital.*, 1956, Nov., 417–421) The principles of solidification in blind feeder heads are set out, and suitable dimensions and methods of calculating them are given. The dimensions for Washburn cores are also given.

The influence of mould material condition and some other factors on the surface roughness of G–AlSi–12 and cast iron W. Patterson and P. Schneider (*Giessereitechn., Wiss. Beih.*, 1959, (26), Oct., 1429–1453) Experiments carried out with a recording microgeometer (Hahn and Kolb, Stuttgart) and a pneumatic method, details of which are given, are described, and the results discussed. A linear relationship was found between gas permeability of moulding sand, and grain size and uniformity, and surface roughness. The effects of additions to the sand and resultant roughness are enumerated, and the influence of surface tension and of metallostatic pressure on roughness are among the factors examined (46 refs).

Green sand principles controlling casting quality R. W. Heine, E. H. King, and J. S. Schumacher (*Inst. Brit. Found. Australian Branch*, 1959, **10,** 109–137) The common defects occurring in sand castings are discussed, followed by consideration of the clay–sand–water aggregate, the use of additives, the effect of moulding on the sand properties, the green and dry compressive strengths of the sand and the avoidance of defects.

The combined influence of shrinkage and atmospheric pressure in the formation of casting defects F. Boussard (*Fonderie*, 1960, Jan., 3–10) The principle of atmospheric pressure with regard to casting is discussed. The part played by various sands and the influence of the position of the casting are dealt with. Correct design, casting temp., section size, and composition of melt and correct feeding are considered of importance in preventing defects.—B.P.

Are the scabs in natural sand moulds different from those in synthetic sand moulds? U. Harada and K. Nishiyama (*Imono*, 1959, **31,** Aug., 724–732) Tests showed differences in the defects in grey iron poured at 1320°C into natural or synthetic green sand moulds.

Pinholing and hydrogen pick-up in malleable cast iron J. V. Dawson (*BCIRA J.*, 1959, **7,** Oct., 824–831) The relationship between metal composition, H_2 pick-up from the mould and pinholing was studied. H_2 pickup is increased by increasing Mn or Al contents, or S in the presence of Al. To avoid pinholing, B should be used instead of Al for malleable cast iron.

Sinking under bosses on thin plates: The effect of variations in pouring temperature A. G. Fuller (*BCIRA J.*, 1959, **7,** Oct., 800–812) The author reports the results of investigations of the influence of variations in pouring temp. upon sinking under bosses on thin plates. It was found that two main factors affect sinking; the number of eutectic cells growing from an iron of a given degree of nucleation and the relationship between pouring temp. and mould cavity. Sinking is at a minimum when the pouring temp. is between 1280 and 1330°C.—T.G.

Hot tears in steel castings M. A. Notte (*Mod. Castings*, 1960, **37,** Feb., 35–37) The various causes of hot-tearing in castings, and the available remedies, both in design and casting, are briefly reviewed.

Measurement vs calculation of solidification of metal in iron moulds J. D. Keller and N. R. Arant (*Mod. Castings*, 1960, **37,** Feb., 71–80) Experimental work was carried out on production rolls of alloy cast iron, of three different dia., in order to obtain temp. measurements. These rolls were of the double-poured type. A new method of calculating temp. at various points in the castings was evolved, and it was shown to give reasonably good agreement with the experimental results.

Solidification of steel in permanent and sand moulds L. Šmrha (*Hutn. Listy*, 1959, **14,** (12), 1039–1048) Solidification was studied in the case of a medium-carbon steel cast into cylindrical permanent or sand moulds; castings weighed about 80 lb. The effect of preheating times, temp., and other variables on the rate of advance of the solidification front, mould-ingot interface temp., and on the thermal gradients in the casting and the moulds, were investigated. The results are generalized as far as possible by the use of appropriate equations, and their import for practical foundry work is discussed.—P.F.

The theoretical bases for determining the mode of solidification of steel castings L. Šmrha (*Hutn. Listy*, 1959, **14,** (11), 936–942) Methods for the calculation of the solidification contours and heat transfer in castings are explained with special reference to cylinders and plates.

Gases in metals in the foundry J. Prentice and R. S. Moore (*Inst. Brit. Found., Australian Branch*, 1959, **10,** 25–34) The origin, appearance, detection, and prevention of gas defects in castings are briefly reviewed.

To make a ton of castings . . . How many tons of materials? . . . and how far do they move? J. H. Schaum (*Mod. Castings*, 1960, **37,** April, 24–29) In order to answer the above questions, the production of castings at the Ewart Factory of Link-Belt Co.—one of the leading producers of foundry handling equipment—was analysed. A detailed account of the analysis is given, and it was found that almost 90 t of material had to be moved during the manufacture of 1 t of finished castings.

VACUUM METALLURGY

World's largest vacuum induction furnace J. W. Byrne (*Met. Prog.*, 1960, **77,** April, 83 86) The unit described is a 5000-lb furnace installed at the works of the Metals Div., of Kelsey-Hayes Co., New Hartford, NY. Features of the equipment are that crucible and tank are a single unit, and can be tilted together, and a multi-phase stirring unit is fitted.

Vacuum melting for the new metals at Jessop-Saville S. E. Schattmann (*Prec. Met. Mold.*, 1960, **18,** March, 44–46, 79) Accounts of the Wild-Barfield vacuum induction and the Heraeus Consumable arc melting furnaces are given and the value of vacuum processes and the structure of low-H_2 steel are considered.

Studies on the vacuum casting process. I. On the construction of large-scale vacuum casting equipment T. Takeiri, Y. Fujimoto, M. Kadose, and J. Watanabe (*Tetsu to Hagane*, 1959, **45,** Sept., 961–962).—K.E.J.

Comparison of vacuum- and air-cast steels. II. Influence of vacuum casting on the quality of steel T. Suzuki and T. Asakuma (*Tetsu to Hagane*, 1959, **45,** Oct., 1145–1151) Twenty months' experience showed the following advantages of vacuum cast steel: elimination of 60%H_2 and 30%O_2, improvement in elongation and reduction of area, 30% reduction of non-metallic inclusions; reduction of secondary cavities, freedom from flakes and shorter heat-treatment cycles.

Some experiments on the degassing effects in vacuum induction melting M. Kawahata, K. Yokota, and T. Watanabe (*Tetsu to Hagane*, 1959, **45,** Sept., 962–964) Graphite additions have a marked effect in reducing the final O_2 content of iron and nickel-base alloy, while H_2 treatment has a less marked effect. The effects of N_2 degassing and degassing times are examined.—K.E.J.

The effect of a vacuum on the changes in the critical temperature, structure and chemical composition of iron V. P. Chernobrovkin, V. S. Belyaev, and A. A. Dobryden' (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 747–751) When melting iron *in vacuo*, its chemical composition changes; Mn and S contents diminish and the proportion of graphite increases. When cooling the iron *in vacuo*, the eutectic transformation occurs at lower and the eutectoid at higher temp. The pearlitic structure of sulphur iron becomes ferritic after remelting *in vacuo* and that of white iron pearlitic-ferritic.

Investigation of the process of the production of chromium metal in vacuo G. V. Karsanov, A. N. Tirkina, and L. S. Odoevskii (*Stal'*, 1960, (4), 321–327).

A penny a pound knocks the H_2 out of steel (*Mod. Castings*, 1959, **36,** Nov., 45) A system of vacuum degassing with simultaneous injection of He can reduce the H_2 content of a 30-t electric furnace melt from 3·9 ppm to 1·5 ppm in 12 min.—A.D.H.

Brazing in vacuum cuts costs, improves quality R. G. Gibbons and A. Goudreau (*Metalw. Prod.*, 1960, **104,** Jan. 20, 115–116) a new type of fast-cycle brazing furnace installed by Utica Division of Bendix Aviation Corporation gives high joint quality and lower process costs in joining complex stainless steel assemblies.

New melting practice for the production of ball bearing steel G. N. Oiks, P. P. Matevosyan, G. A. Sokolov, I. I. Ansheles, V. I. Danilin, and B. Z. Kononov (*Stal'*, 1960, (4), 308–313) Vacuum treatment in the ladle had no effect on inclusions. Vacuum treatment in the furnace followed by deoxidation with Si and Al in the ladle was necessary.

REHEATING FURNACES AND SOAKING PITS

Bibliography on soaking pit practice 1946–1959 (*Bibliography No.35*, 1959, pp.2) With special reference to modern one-way fired pits.

The planning of reheating furnaces for high-throughput finishing mills H. J. Pohle (*Stahl Eisen*, 1960, **80,** May 12, 670–675) Design, construction, operation, and control equipment of newly-built reheating furnaces for a finishing mill for light sections and strip are described.—T.G.

Modern design of slab reheating furnaces H. C. Henschen (*Iron Steel Eng.*, 1959, **36,** Oct., 75–82) Design features, and operating characteristics are described of three types of modified slab reheating furnace—3-zone, 4-zone, and 5-zone, at the Sparrows Point plant of Bethlehem Steel Co.

The correct number of soaking pits for blooming mills S. E. Rostkovskii (*Stal'*, 1960, (3), 274–278) A queueing theory treatment of operations in which pits act also as ingot stores to smooth out irregular working is discussed.

Modern regenerative soaking pits H. S. Hall (*Iron Steel Eng.*, 1959, **36,** Oct., 83–92) The construction and control design of several types of regenerative soaking pits are described. It is shown that suitably designed soakers can be used with a large range of fuels.

Experience of converting the furnaces of the sheet-rolling shop to natural gas I. S. Zhilin (*Stal'*, 1960, (3), 278–279) Conversion of the 'Hammer and Sickle' works in furnaces of various types was carried out in 1956 but improvement in burner design has continued and these are described. Output increased and some equipment was simplified with considerthese are described. Output increased and able reductions in cost.

Furnace improvements increase rod mill capacity D. W. McLean and K. A. Gmell (*Iron Steel Eng.*, 1960, **37,** Jan., 77–84) A description of the re-designing of a billet heating unit to increase its productive capacity from 25 to 59 t/h without increasing its overall dimensions.

Waste heat recovery in metallurgical furnaces. III I. Stoicescu and C. Stamatescu (*Met. Constr. Masini*, 1959, (12), 1038–1042) Heat recovery from soaking pits and forge furnaces is considered, also from cupolas. The problem is set out in the form of suggestions and further comments are invited from interested specialists.

Influence of sulphur in the heavy oil used for an ingot-reheating furnace on the working face of the rolled material T. Mori and M. Kamakura (*Tetsu to Hagane*, 1959, **45,** Sept., 979–981) Microstructures of high- and low-sulphur scales are described.—K.E.J.

Catalyst applications to continuous strip ovens R. J. Ruff (*Wire Wire Prod.*, 1959, **34,** Oct., 1346–1348, 1390–1391) The use of catalytic processes to oxidize contaminants in furnace exhaust gases and supply heat for recovery to the furnace or for outside heat requirements is described and illustrated with examples.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Industrial furnaces H. Schwiedessen (*Brennstoff-Wärme-Kraft*, 1960, **12,** April, 167–168) This is a review of developments reported in the literature during 1959. The outstanding development during 1959 was in the control sector with a view to automation (55 refs).

Modern continuous furnaces for heat treatment F. Rüb (*Europäischer Masch. Markt.*, 1960, **10,** (2), 21–24) An illustrated review of small part, pusher, passage, conveyor, and roller hearth furnaces and temp. control.

On the measurement of heat transfer in metallurgical furnaces L. Tomis (*Hutnik Listy*, 1959, **14,** (11), 947–950) The performance, advantages, and design of a new water-cooled differential calorimeter for the measurement of temp. fluctuations in metallurgical furnaces are described.—P.F.

Report on the conference on thermal treatments A. Clerc (*Centre Doc. Sidér. Circ.*, 1959, **16,** (4), 801–807) Brief abstracts are given of papers presented at the conference held at Wiesbaden, Oct. 1958. Subjects dealt with included factors influencing hardenability of case-hardening steels, and other theoretical and practical aspects of case hardening; use of thermoelectric potential in non-destructive testing, i.e. in measuring case depth and C content; causes and possible modification in banded structures in steel; C precipitation from martensite and formation of carbides during heat treatment, in 1%C, low-alloy steels; maintenance of hardening baths; and bend tests on tool steels. **Report of the international conference on the influence of thermal treatment on resistance to fatigue** (809–812) Subjects dealt with were the removal of residual stresses after thermal treatment from plain C and alloy steels; the relations between thermal treatment, distribution of stresses,

and fatigue resistance after certain surface heat treatments; and the increase of fatigue strength obtained after bath nitriding.

Report of the third International Conference at Ilmenau (East Germany) A. Clerc (*Centre Doc. Sidér. Circ.*, 1959, **16**, (4), 813–817) Papers abstracted briefly dealt with temp. regulation in a furnace installation, modern thermal treatments, surface hardening techniques, and recrystallization.

Recuperative town gas-fired furnaces for temperatures up to 2000°C E. A. K. Patrick and R. D. Hastie (*Coke Gas*, 1959, **21**, Dec., 507–511) The working chamber is 12 × 12 × 6 in.

The radiant-heat flow in the hearth of the combustion-type furnace Jen Shih-cheng (*Acta Met. Sin.*, 1959, Sept., **4**, 217–221) [In Chinese].—K.E.J.

Which bell furnace? L. W. Johnson (*Met. Prog.*, 1960, **77**, March, 87–91) Features of the cold retort, hot retort, and radiation shield types of furnace are described and compared, and some applications indicated.

Practical aspects of atmosphere control J. Perrine (*Iron Steel Eng.*, 1960, **37**, Feb., 121–131) Problems connected with the properties of various protective atmospheres are discussed. N_2 pickup of stainless steels in dissociated NH_3 is examined.

Getting the most out of your endothermic generator (*Efco J.*, 1960, **1**, Jan., 17–20) Operating instructions are given for the equipment.

Controlled furnace atmosphere in heat treatment of steel A. L. Forrester (*Gas World*, 1960, **151**, April 16, 73–74, 76, 78) A general account of atmospheres and atmosphere generators, their operation, control, and analysis.

On the measurement of carbon potential in the furnace atmosphere A. Adachi, T. Banno, and T. Enami (*Tetsu to Hagane*, 1959, **45**, Sept., 1123–1125) Details are given of a test method depending on the relationship between carburized carbon % and the increase of resistance of an iron wire.—K.E.J.

Cooling conditions for cast metals in high-chromium steels N. M. Lapotyshkin (*Stal'*, 1960, (3), 259–263) Internal cold cracking in transformer steel was studied by picking up the sound emitted on crack formation by a device described. Reduction of Si reduces the amount of cracking and also the temp. at which it occurs. Cooling conditions minimizing crack formation were worked out.

Effect of cyclic heat treatment on dimensional changes of aluminium and Armco iron specimens G. P. Lazarev (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 57–58) Changes in the dimensions of these specimens caused by cyclic heat treatment depend on the type of crystal lattice also on the ratio of length to dia. (also thickness) of specimens. A considerable variation in this ratio may result in a change of sign of these dimensional variations.

Automatic hardening and tempering of carbine barrel forgings G. W. B. Furnaces Ltd (*Metallurgia*, 1960, **61**, Jan., 32–33).

The earliest days of precipitation hardening W. A. Mudge (*Met. Prog.*, 1960, **77**, April, 79–82) An historical survey (31 refs).

Spark-hardening of metals N. C. Welsh (*J. Inst. Met.*, 1959, **88**, Nov., 103–111) Steel hardening was studied and atmospheric N_2 and C from the oil were shown to promote hardening.

Construction of an electrolytic hardening machine T. Kuroda and H. Kondo (*Nippon Kinzoku*, 1959, **23**, March, 164–168) A trial machine of the dipping type, of 20 kW capacity, is specially designed for equality of heating and modulation of quenching speed. Full details are given.—K.E.J.

Versatility feature of gas carburizing unit T. W. Ruffle (*Metalw. Prodn.*, 1960, **104**, Jan. 6, 9–11) Arrangements for heat-treatment of heavy-duty transmission units with multiple two-zone furnaces and automatic transfer are described.

Investigating high-temperature cementation of steel in a solid carburizer with high frequency current heating S. F. Yur'ev and S. K. Ginzburg (*Metalloved. Term. Obra. Met.*, 1959, (8), 32–38) An increase in the temp. of cementation in a solid carburizer raises the efficiency of the process at 1000° by a factor of 3–4 and at 1150° by one of 5–8 as compared with the normal procedure. The main advantages in the use of h–f current is the prevention of local carburizing at points of contact with the carburizing particles. A twofold heat treatment should follow so as to impart high mechanical properties to the cemented layer.

Oxygen and hydrogen diffusion in steel during cementation O. K. Kotov (*Metalloved. Term. Obra. Met.*, 1959, (8), 43–45) It has been experimentally established that during the cementation of leaves or blades of springs with spindle oil and solid carburizer, steels become saturated with both O_2 and H_2. If triethanolamine is used, the surface layer is also saturated with N_2.

Modern practice of flame hardening M. R. Scott (*Met. Prog.*, 1960, **77**, April, 69–73) An outline of the three methods available: progressive, spinning, and combination, and the materials which can be treated.

Case histories of flame hardening applications (*Met. Prog.*, 1960, **77**, April, 74–75) Illustrations and brief notes on several large flame-hardened parts.

The flame hardening of gears J. T. Howat (*Met. Prog.*, 1960, **77**, April, 76–78) Equipment, costs, and production costs are discussed, and a table of alloy steels for flame hardening is reproduced, with hardness values.

Temperature control during induction heating J. F. Libsch (*Met. Prog.*, 1960, **77**, March, 94–98) Some systems of rapid-action controls to monitor induction heating processes are described, with outline circuit diagrams. The systems are based on the use of the signal from thermocouple or radiation elements, or current proportioning control.

Contribution to the clearing up of the problem of distortion in the flame- and induction-hardening of bulky grey iron castings K. Falkenmayer (*Hart. Techn. Mitt.*, **13**, (3), 133–199) The experiments described investigate the cause and types of distortion, individual factors influencing their occurrence, and possible measures to be taken to diminish resultant distortion (37 refs).

Special features of the nitriding of 1Kh13 steel I. V. Samolov (*Metalloved. Term. Obra. Met.*, 1959, (8), 39–41) To avoid chipping of the nitrided diffusion layer, this steel should contain not less than 0·1%C and not over 13%Cr. A thin copper layer will give a high quality nitrided skin. Best results are obtained with a double layer of copper and tin.

Furnaces and equipment for nitriding and carbo-nitriding steel P. J. Kool and J. van der Linde (*Metalen*, 1960, **15**, April 15, 115–119) The author discusses with the help of numerous illustrations the different types of furnaces and equipment available on the market for the heat treatment by nitriding and carbonitriding of steel using various gases and salts in order to impart skin hardness.—F.R.H.

On the sulphurization of cast iron M. Shiozawa, H. Nakai, and Y. Takeuchi (*Tetsu to Hagane*, 1959, **45**, Sept., 909–911) The effects of time, temp., and iron composition on increase in wt. by sulphurization and the composition of the sulphurized outer layer are described quantitatively.—K.E.J.

Measurement of decarburization in alloy steels G. Meikle (*Metallurgist*, 1960, **1**, March, 65) The value of the heat-treatment method of Payson and Savage is confirmed.

Some recent heat treatment developments J. McMullen (*EFCO J.*, 1960, (1), Jan., 2–9, 12) An illustrated review of annealing and hardening processes.

Investigating the heating and cooling of coils in covered MMK furnaces E. G. Dveirin (*Stal'*, 1960, (4), 366–370) Studies of three-stand bright annealing coil furnaces in the MMK works tinning shop were made. Coil dimensions, temp. distribution, and the circulation of the protective atmosphere were taken into account and various experimental conditions were used.

Bright annealing of stainless steel tubes T. Okamoto and K. Takahashi (*Tetsu to Hagane*, 1959, **45**, Sept., 1008–1010).—K.E.J.

Electric furnaces solve new heating problems J. E. Friess (*Metalw. Prodn.*, 1959, **103**, Dec. 4, 1962–1964) Examples from the U S of controlled atmosphere or vacuum annealing installations.

A conveyor belt furnace for annealing centrifugally cast pipe A. V. Skvortsov (*Stal'*, 1960, (4), 375–378) A description of a five-zone furnace with some automatic units is given. The heating is carried out with blast-furnace gas and the furnace holds 80 pipes, 100 mm dia. and 150 mm dia. weighing 9·7 t which pass through in 40 min. or less. Modifications to the design are suggested.

Working out the process for annealing cold-rolled dynamo steel sheet in continuous furnaces N. F. Dubrov (*Stal'*, 1960, (4), 370) A research note. The amount of decarburizing necessary is studied and it is shown that intermediate annealing in moist H_2 is effective.

Modern methods of wire annealing F. Frodsham (*Draht*, 1960, **11**, April, 167–170) The fundamentals of the heat-treatment of steel and in particular, of steel wire are reviewed and annealing furnaces such as salt-bath furnaces, muffle furnaces heated by gas or electricity, with and without a protective atmosphere are described.—T.G.

Study on high speed tools. XIX. Stepped annealing of high speed steels (SKH8 and SKH6) H. Hotta (*Tetsu to Hagane*, 1959, **45**, Nov., 1272–1275) This rapid softening method was studied. The hardness of quenched high-speed steel annealed from 810° to 800°C by air, water, or oil cooling was nearly equal to that obtained in a full annealing furnace (21 refs).

Continuous annealing in molten sodium J. B. Keller (*Iron Steel Eng.*, 1959, **36**, Nov., 125–133) Advantages of the techniques are low capital cost and compactness of the plant and high production rates, compared with conventional continuous annealing plant for steel strip. Design and problems encountered in development are discussed, including questions of health and safety.

Improving the quality of wires for reinforced concrete E. E. Anastasiu and G. Socolescu (*Met. Constr. Masini.*, 1959, (7), 580–585) By heating the 5 mm dia. wires made by the Industria Sirmei of Cimpia Turzei, for 30 sec. at 350°, an extra elongation of 2% is obtained The treatment is easy to apply and does not appreciably increase production costs.

Austolize or Austulize P. Payson (*Met. Prog.*, 1960, **77**, April, 132) A letter. Discussion of a suitable term to define 'Austenitizing temperature'.

Prevention of surface defects when annealing tinplate K. V. Dneprenko (*Stal'*, 1960, (3), 255–259) Spots and dark edges not removed in pickling can be prevented by the careful exclusion of O_2 from the protective atmosphere in annealing. The O_2 can be removed by increasing generator temp.

Structural transformations after the annealing of previously plastically deformed s.g. iron I. Cornea and S. Zamirca (*Revue Roumaine Met.*, 1959, **4**, (2), 213–216) [In French] The specimens were subjected to four degrees of reduction 32, 53, 63, and 79% and after annealing at 600, 680, and 760° for periods between 3 and 30 h, the first change observed was the appearance of ferrite in different proportions corresponding to the temp. and holding period. Under the microscope it was observed that the unchanged pearlite regions were concentrated especially around the P eutectic. In specimens annealed for long periods a new component was observed, component 'X" described by Tabadze. A new component termed by the authors 'X' was also observed which appears in quantities which vary directly with the temp. and the duration of the anneal. Below 6 h at 680°, this component, with granular appearance, seems to have been partially replaced by graphite.

Heat treatment of stamped components F. I. Golomazov (*Metalloved. Term. Obra. Met.*, 1959, (8), 51) In certain plants, heat treatment of high carbon steel components in hot media has resulted in increased hardness (mixtures containing KNO_3 of low m.p.). A typical mixture with a m.p. of 110° for quenching steel is given as follows: 7%$NaNO_3$, 53% KNO_3, 40%$NaNO_2$, and 5% water.

On some experiments of the quenching of low-carbon steel S. Yonetani (*Nippon Kinzoku*, 1959, **23**, March, 137–141) Work is reported on 0·13–0·2%C, 0·7–2%Mn steel for use in large chains. The changes of mechanical properties

by heat treatment are related to the growth of γ-grains in the austenite region. The rate of ferrite pptn. at the subcritical temp. and its hardness after quenching are closely related to the state of Al, N, and AlN dissolved or pptd. in the γ-region.—K.E.J.

Sub-zero treatment of quenched steel. II. Effect of the addition of small amount of elements on the stabilization of retained austenite Y. Imai and M. Izumiyama (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1959, **A, 11,** Oct., 393–400) [In English] Stabilization after partial transformation to martensite was investigated using 1%C steels containing small amounts of Al, Ti, Ni, or Sn, and an Fe–Ni alloy. It was difficult to stabilize the austenite with steels containing Al and Ti, but easy with those containing only Si and Mn, Ni, or Sn. No stabilization was observed on Fe–30%Ni alloy. Stabilization is probably caused by interstitial solute atoms, e.g. C and N (18 refs).—K.E.J.

Sub-zero treatment of quenched steel. V. Stabilization of austenite martensite transformation Y. Imai and M. Izumiyama (*Nippon Kinzoku*, 1959, **23,** March, 156–160) In 1·12%C 4·98%Ni steel, the stabilization degree (expressed by δ, the temp. lag of the transformation point) is mathematically related to the ageing time at room temp. The correlation between δ and the % of stabilized austenite is not linear. Stabilization proceeds rapidly when the previously transformed martensite increases beyond 65%.—K.E.J.

Ipsen scaled quench unit for automatic heat treatment Ipsen Industries Inc. (*Ind. Gas.*, 1959, **22,** Nov., 12–16) The furnace is described and illustrated. The carburizing performance of the installation is given.

Quick die quench shatters forming bottleneck Ryan Aeronautical Co. (*Steel*, 1960, **146,** Feb. 15, 134) The use of a quench die cooled with cold air is described. The air is cooled with dry ice and then blown through passages in the die. Thus the die is kept from heating up and can be used continuously.—D.L.C.P.

Quenching carbon steel tools in fused salts B. G. Gold'shtein and A. I. Vinogradov (*Metalloved. Obra. Met.*, 1959, (7), 45–49) A mixture of 53%KNO_3+40%$NaNO_2$+7%$NaNO_3$ with 2·3·5% water has a m.p. of about 100° and without water of about 140°. This salt mixture makes it possible to carry out a gradual quenching at low enough temp. (110–125°) and with intensive mixing and a little water gives the cooling conditions necessary to obtain the required hardness for carbon tool steels in thicknesses up to 60–80 mm. More water reduces the m.p. still further so that quenching can be carried out at even lower temp. with this mixture.

Low heat tempering of cemented and quenched 30KhGT steel V. V. Babayan (*Metalloved. Term. Obra. Met.*, 1959, (8), 41–43) The optimum heat for tempering this steel is between 180 and 200° with a holding time of 90 min. The impact strength of the cemented layer 1·1–1·2 mm) between 120–150° is lower by 20–40% than with the optimum temp. and that of the non-tempered cemented layer is lower by 50%.

Changes in mechanical properties during the fourth-stage tempering of a chromium steel C. Bodsworth *et al.* (*JISI*, 1960, **196,** Sept., 57–59) [This issue].

Study on the behaviour of retained austenite for alloy white cast iron as tempered Y. Sensaki and I. Doi (*Tetsu to Hagane*, 1959, **45,** Sept., 905–907) Details are given for differential dilation, hardness, microstructure, etc.

FORGING, STAMPING, DRAWING AND PRESSING

Forms and shapes of materials (*Mat. Design Eng.*, 1959, **50,** *Materials Selection Issue* 1959–60, 341–373) Tables of methods of production and fabrication showing suitable materials, size, properties, and precisions are shown (342–343). More detail is then tabulated for castings (344–347), extrusions (348–350), stampings and pressings (351–356), forgings (357–359), tubing (364–367), and wire (368).

Forging line mechanized from steel store G. H. De Groat (*Metalw. Prod.*, 1959, **103,** Oct. 16, 1648–1650) Mechanized lines installed at Saginaw Steering Gear Division of General Motors Corp. to forge automobile tailshaft parts are described.

A C-stand forging press H. J. Pahnke (*Werkstattstechn.*, 1959, **49,** Oct., 649–650) The design and operation of the press are described.

How to produce forging dies by cold and hot working P. Grüner (*Europäischer Masch. Markt.*, 1960, **10,** (2), 9–14) The processes are described with discussion of die design and showing structures. Tables of composition, heat-treatment, and properties are included.

New drop-forging steels for hot deformation G. A. Khasin and G. I. Parabina (*Stal'*, 1960, (4), 354–357) Studies on four newly-developed and six imported steels were made and comparisons with 5 KhNV and 3 Kh2V8 are reported. The properties of the steels are set out with compositions and other data and recommendations are made.

Comparative efficiency of American and German die steels H. Hecht and H. M. Hiller (*Werkstattstechn.*, 1959, **49,** Oct., 645–648) Properties of a series of American die steels, i.e. composition and heat-treatment behaviour, were examined to ascertain their range of applicability, particularly in relation to German practice.

The work of mechanical deformation in die forging C. Gazdaru (*Met. Constr. Masini.*, 1959, (7), 576–579) The method of calculating the mechanical work is based on the assumption of a single deformation. In practice, deformation is not uniform either in extent or in direction. These factors are considered in the calculations submitted by means of correction coefficients experimentally established. Results obtained have given satisfactory approximations.

The evaluation of possible alternative forging techniques J. Banbury and J. V. Chelsom *BISRA Report List No.*103, July–Aug., 1959, pp.7) Preliminary studies by the Operational Research Dept. on the possibilities of mechanization and the reduction of the skilled labour required are reported.

Forgeability of ferritic steels for superheater tubes O. Bihet and M. Caubo (*Hutn. Listy*, 1959, **14,** (12), 1103–1105) A study of the effect of additions of V, Nb, Ti, and Co steels containing 9%Cr and 1%Mo, with about 0·1%C, is described, mainly with regard to the ease of rolling the tubes. The carbide forming elements were added to increase the creep resistance. The workability, as determined in torsion tests, increases with the ferrite content of the steels.—P.F.

Determination of forging factors for improving mechanical properties: on the properties of forgings T. Momose, T. Izu, K. Oshima, and A. Yoshida (*Tetsu to Hagane*, 1959, **45,** Sept., 975–978) The effects of rolling ratio, forging ratio, heating temp., and heat treatment on mechanical properties are evaluated statistically.—K.E.J.

On the centre-line temperature of large steel forgings during the forging process K. Kawano and R. Yanagimoto (*Tetsu to Hagane*, 1959, **45,** Sept., 974–975) Surface and centre-line temp. are plotted for up to 150 min. after discharge from the furnace, for forgings of dia. in the range 300–800 mm.—K.E.J.

Surface hardening of drop-forging dies J. Elfmark and J. Štěno (*Hutn. Listy*, 1959, **14,** (12), 1060–1064) Methods of surface hardening by the use of acetylene burners were developed for use with drop-forging dies employed in the manufacture of balls for mills. The method resulted in a five fold increase in die life, and also gave good results in hardening piercing tools used for punching holes in railway sleepers.—P.F.

Three-broach set-up shapes pole-pieces W. H. Marley & Co. Ltd (*Metalw. Prod.*, 1960, **104,** Jan., 20 109–110) A broaching fixture incorporated on a Marlco 12-t hydraulic press gives high dimensional stability to small parts.

Flying press in warehouse helps small companies R. E. Hasselbring (*Steel*, 1960, **146,** Feb. 15, 124–125) A stamping press made by the Wean Equipment Co. is mentioned without any details.

Stronger parts from work-hardened sheet-metal H. A. Jahnle (*Prod. Eng.*, 1959, **30,** Oct. 26, 76–78) An account in general terms with notes on welding.

How control of burrs aids sheet metal stamping F. Strasser (*Iron Age*, 1960, **185,** Jan. 21, 90–92) Advice is given on means of minimizing burrs on metal stampings. This gives benefits in maintenance of equipment as well as improving quality of parts.—D.L.C.P.

The elastic behaviour of wide presses H. Neumann (*Werkstattstechn.*, 1959, **49,** Oct., 619–624).

Overhead handling cuts press shop overheads J. R. Couchois (*Metalw. Prodn.*, 1960, **104,** April 6, 608–609) An account of Convair-Astronautics practice, with illustrations.

Development of rapid hot-forming presses F. Meyercordt (*Draht*, 1959, **10,** Sept., 475–480) Examples are given of modern equipment for the manufacture of bolts and similar parts at a high rate of production. Operational and economic factors are considered.

Finish piercing (*PERA Bull.*, 1960, **13,** March, 54–62) An outline of Report No.70. Punch shape, piercing force, burr height, and hole taper, shear mechanism, and tapping torque were studied for hot and cold rolled steels and for Al and Cu.

Chain saw cutter bars are a press product M. Conklin (*Canad. Metalw.*, 1960, **23,** March, 28–29) Punching, spot-welding, and heat-treatment are described.

Steel for strip intended for stamping and deep drawing E. Diaz-Varela and C. Vivas (*Inst. Hierro Acero*, 1959, **12,** (59), 1–27) The properties and structures of steels required are examined, the object of a skin pass and ageing, and the resulting properties of the strip are discussed, and some notes are given on the manufacture of the steel (23 refs).

High-energy rate forming of metals H. P. Tardif (*Canad. Metalw.*, 1960, **23,** April, 24–28, 53) A general account of explosive forming with illustrations of parts.

Report on explosive forming. I. Theory and practice of high-energy forming R. A. Cooley (*Met. Prod.*, 1959, **103,** Oct. 16, 1645–1647) An outline of the three techniques available, and some applications.

Du Pont reports an explosive forming (*Steel*, 1959, **145,** Nov. 23, 74–76) Experience of explosive forming is surveyed. The process may be considered for single shot forming, finish sizing, improved mating of parts, hardening of austenitic steels. Almost any material with appreciable ductility can be formed. Examples of parts made by the method include: an 8 ft dia. hemisphere in ⅛ in. steel for tank-ends; cup and cone shaped parts to great dimensional accuracy.—D.L.C.P.

Kaiser instals large plate stretcher at Ravenswood Works (*Iron Steel Eng.*, 1960, **37,** Feb., 145–146) This stretcher has a tractive force of 30×10^6 lb, and was built for Kaiser Aluminium Co. by Hydraulik G.m.b.H. of Duisburg.

Definitions and methods of calculation for tube drawing A. Mohrnheim (*Draht*, 1959, **10,** Oct., 517–522) Processes of tube drawing are defined as in 'Metals Handbook' 1948, and simple mathametical relationships are derived, and illustrated by examples, for calculating tool dimensions, plastic deformation, etc. The use of special slide rules is described (17 refs).

Cooling of wire drawing equipment H. Sedlaczek and K. Becker (*VDIZ*, 1959, **101,** Oct. 21, 1385–1389) Problems of cooling modern high-speed equipment and the various systems employed are discussed.

The drawing and handling of long production run wire coils U. V. Johnson (*Wire Wire Prod.*, 1959, **34,** Oct., 1281–1283, 1402–1405) Development work carried out by the Portsmouth Division rod and wire mills of Detroit Steel Corp. in increasing the weight of wire coils produced is described. This work, commenced in 1952, culminated in 1957 in the installation and modification of equipment to produce coils in bundles up to 4000 lb. This equipment and the procedures adopted are described.

Diameter measurement on moving wire G. Sachs (*Draht*, 1959, **10,** Sept., 465–471) Various systems of measurement, evaluation, and recording, with and without contact, are described and illustrated.

Tube bending controlled by tape Baldwin-Lima-Hamilton Corp. (*Iron Age*, 1960, **185,** Jan. 12, 94–95) The operation of, and a use for the 'Bend-O-Matic' machine which bends thin

wall pipe, are described. The machine automatically bends tubing, according to information, fed into an electronic control system on punched tape. The application described is the production of car exhaust pipes. Distributors can make patterns to order instead of holding a large variety of finished stocks; the possible savings are evaluated.—D.L.C.P.

Select bibliography on bending, straightening and reeling of steel 1940–1959 Sheffield City Libraries (*Res. Bibliography* 42, 1959, Aug., pp.12).

Chipless forming (*Draht*, 1959, **10,** Oct., 524–527) Several examples of the manufacture of special parts by cold and hot forming are given to illustrate the successful use of rationalization.

Metal spinning (*Met. Prog.*, 1959, **76,** Oct., 118–121) Metal spinning, defined as the pressure forming of metal on a rotating chuck or die, is described in its modifications chuck, displacement and power spinning. Design considerations are outlined.

The spinning of steel J. Lomas (*Mach. Lloyd*, 1960, **32,** Jan. 30, 41–43) A description of the process in general terms.

Hot extrusion A. B. Tesmen (*Met. Prog.*, 1959, **76,** Oct., 114–118) Size limitations, section thickness and tolerances, and various modifications of the technique are outlined briefly and illustrated by examples.

Cold extrusion E. J. Epner (*Met. Prog.*, 1959, **76,** Oct., 122–125) Advantages of the cold-extrusion process for small parts are discussed, and considerations governing design are reviewed with examples.

Cold extrude stainless shafts R. H. Eshelman (*Iron Age*, 1960, **185,** Jan. 21, 86–87) Propellor and drive shafts for outboard motors are cited as an example of a suitable application for the cold extrusion of stainless steel. Bar stock is extruded through carbide dies. Layers of oxalate, and a soap-type material are used as lubricants. The technique is warranted by the following factors: the part is of round section; saving of steel is effected; machining is reduced; demand for the part justifies cost of tools etc.—D.L.C.P.

Forming cold alloy steel by impact extrusion Matthews Manf. Co. (*Weld. Met. Fabr.*, 1960, **28,** Feb., 60) Tolerances and subsequent heat treatment as well as the yield strength of the completed parts are briefly given.

Extrusion of ferrous metals J. H. Rice (*Prec. Met. Mold.*, 1960, **18,** March, 48–49) A general survey with notes on possible developments and a diagram of extruded sections.

Requirements for machine tools for extrusion pressing H. D. Feldmann (*Werkstattstechn.*, 1959, **49,** Aug., 421–429) Various types of presses and their fields of application are reviewed.

ROLLING MILL PRACTICE

Rolling mill practice A. Wolf (*VDIZ*, 1959, **101,** Oct. 21, 1409–1414) A review (154 refs).

Metal pressure on the rolls and the temperature conditions for hot rolling steel-tombac bimetal batches I. Ya. Ryazantsev (*Tsvet. Met.*, 1959, (12), 55–60) Rolling schemes which make use of the full possibilities of the mill for every pass are not possible in view of the overloading of the motor. Only in the first two passes can rolling be intensified. This would facilitate the conditions for welding the tombac to the steel.

Deformation of the s.g. cast iron in rolling S. Bereu, I. Dragan, and C. Rizescu (*Revue Roumaine Met.*, 1959, **4,** (2), 181–188) [In German] From experiments with specimens rolled with different rates of reduction it may be concluded that the graphite spheroids maintain their unbroken contours up to a reduction of 40%. Beyond this, and preferentially in the direction of rolling, the graphite spheroids become elongated forming very thin strands towards their ends so that the greater the reduction the streakier the spheroids become. Structural changes could also be observed, among which is an enrichment of the matrix in pearlite with increasing deformation.

Comparison of Russian methods of calculating loads in the hot rolling of steels Z. Wusatovski (*Zeszyty Naukowe Pol si Mechanika*, 1956, (3), 19–46; *TMS* 366, pp.33) A comparison of seven methods of calculation is presented. The value of the methods is shown to differ with the type of mill considered.

Semi-hard ductile iron rolls R. Kamenský (*Hutn. Listy*, 1959, **14,** (9), 762–768) Equipment and technique used for the partial inoculation of iron in the production of semi-hard rolls up to 5-t weight are described, and the satisfactory service results obtained are reported (22 refs).

Effect of the number of passes on the spread and elongation of billets in hot rolling Y. Koike, Y. Toda, K. Tihara, T. Kuwahata (*Tetsu to Hagane*, 1959, **45,** Oct., 1152–1157) Square (25–45 mm) bars of low-C steel were rolled in a two-high pinion-drive mill, with rolls of 250 mm dia. With more passes, the total spread decreased and elongation increased. Spread per pass increased with the number of pass. Side barreling was more marked with smaller bars. The max. rolling load was found at the last pass to achieve a given reduction.

The disclosure of hidden production losses in sheet mill shops A. V. Eme'yanov, N. A. Naumenko, and M. Ya. Kal'skii (*Stal'*, 1960, (4), 371–374) A study of production losses due to short interruptions, mainly owing to inexact coordination and process timing is reported. Time studies over some months led to some redesign and retiming and to increased production.

Problems in the manufacture of modern cold rolled transformer sheet steel M. Markuszewicz and J. Groyecki (*Hutnik*, 1960, **27,** (2), 56–65) In investigating the problem of production of cold-rolled transformer plate using the Goss method, permeability and coercion were considered rather than energy losses. It was found that chemical purity of the metal gives a crystallographic structure which leads to satisfactory magnetic properties. The plastic cold working must be carried out correctly, with heat treatment between operations. Furthermore the crystallographic structure before heat treatment was found to be of importance.

Earing of deep-drawn steel sheet parallel and perpendicular to the rolling direction G. Oehler (*Stahl Eisen*, 1960, **80,** April 28, 627) The author reports on earing phenomena occurring in OH steel of deep-drawing quality parallel or perpendicular to the rolling direction. Up to now it was assumed that earing could occur only at an angle of 45° to the rolling direction, but obviously earing can also develop parallel and perpendicular to it. The blanks were obtained from strip. An explanation of this phenomenon cannot yet be offered.—T.G.

Automation (*Colvilles Mag.*, 1959, Summer, 6–9) An account of the fully automatic Dalzell rolling mill.

Automation in processing lines J. J. Remley (*Blast Furn. Steel*, 1960, **48,** Jan., 58–67) The author reviews recent advances in the automation of processing lines, such as developments in strip propulsion equipment, loop controls, power supplies for adjustable voltage drives, silicon rectifiers for cleaning and plating, annealing furnaces, high-speed jet cooling, pinhole detectors, tinplate surface inspection, and automatic data processing system.—T.G.

Profile check at hot mill ensures quality strip D. H. Dellinger (*Steel*, 1960, **146,** Feb. 15, 128) The use of a recording profilometer to check the edge-to-edge surface quality of strip coming from a hot strip mill is described. Periodically a 4 in. length of the strip is cut out, cooled, and tested. The profilometer is run along the edge of the strip and records a continuous trace of the profile. Surface defects, such as ridges due to roll wear, can be detected and remedied immediately.—D.L.C.P.

On the variation of reduction in the temper rolling of tin plate S. Teshima, S. Yasuda, K. Nakajima, and N. Ando (*Tetsu to Hagane*, 1959, **45,** Sept., 987–988) Various operating parameters of a skin-pass mill are interrelated graphically.—K.E.J.

Some experiments on cold strip rolling H. Takahashi and S. Miyoshi (*J. Mech. Lab. Japan*, 1959, **5,** (1), 35–43) After a discussion of the yield stress, some experimental results for cold strip rolling are given, and the results are compared with predictions based on the theories of Sims and Bland and Ford.

Pinch rolling makes cheaper turbine blades (*Metalw. Prod.*, 1960, **104,** Jan. 20, 107–108) The operation of an automatic-cycle, cold rolling machine developed by the de Havilland Engine Co. for precision forming compressor-turbine blades is described.

Tandem cold mill control J. W. Cook (*Blast Fur. Steel Plant*, 1959, **47,** Nov., 1187–1190, 1194–1195) A description of a control system for a modern five-stand tandem cold mill with automatic gauge control. The system of control of the individually-driven stands is explained, and schematic diagrams showing the electrical drive connexions and a tandem mill stand drive control are reproduced.

Forces operating in cold roll forming V. I. Davydov and Yu. K. Puzyrevskii (*Stal', Rolling and Tube Manf. Supp.*, 1959, 95–103) The vectors and pressures involved are identified and formulated, and a simplified calculation of power requirements is presented, with a procedure for approximate calculation.

Studies of rolling in a seamless tubing plug mill. I. Measurements of rolling loads in a large seamless tubing plug mill K. Miyamoto, T. Tanaka, T. Shirai, and Z. Baba (*Tetsu to Hagane*, 1959, **45,** Sept., 984–986).—K.E.J.

Application of speed regulators to the Geneva works hot strip finishing mills G. S. Koss and L. E. Ringger (*Iron Steel Eng.*, 1960, **37,** Jan., 53–63) A description of the design and operation of a speed regulating system to improve dimensional stability of the strip.

On the strain in rails and the straightening forces in the roller-straightening of rails Y. Koike, M. Nishikubo, Y. Toda, and S. Kumagai (*Tetsu to Hagane*, 1959, **45,** Sept., 981–982).

Automatic thickness control—Geneva Works hot strip mill L. E. Ringger and G. S. Koss (*Iron Steel Eng.*, 1960, **37,** Feb., 110–120) Details of the underlying principles and operation of the thickness control system at the Geneva works of United States Steel Corporation; based on automatic screw-down and speed-tension control.

Hydraulic stretchers W. Dohrn and H. Sandherm (*Z. Metallk.*, 1960, **51,** April, 193–198) Hydraulic stretchers are used for stretching, straightening, and detwisting extruded or rolled products. The authors describe the Schloemann stretchers ranging from 25 t to more than 700 t. The use of those machines is illustrated and outlined.—T.G.

Despatch of rolled products by theoretical weight (*Stal'*, 1960, (4), 333–337). **Part I** G. M. Katanel'son (333–334) The work of Burnashov is referred to but his conclusions are challenged. The method of routine measurement of section length at Petrousk is described, but despatch by actual weight is preferred for small sections unless automatic measurement and calculation can be installed. Large and medium sections can be despatched by theoretical weight. **Part II** M. L. Mirenskii (335) The savings from rolling to minus tolerances are noted and the difficulties in measurement are stressed. **Part III** L. Ya. Kats (335–337) Possible methods of accountancy and the question of bonus payments in the application of minus tolerances and despatch by theoretical weight are discussed.

MACHINERY AND SERVICES FOR IRON AND STEEL PLANT

The efficient installation conditions of waste heat boilers I. B. Gol'denburg (*Stal'*, 1960, (3), 285–287) Various factors in the efficient operation of boiler plant, such as the gas-tightness of the pipes and of the reversal gear and regenerators of the OH furnace itself are discussed.

Mobile units reduce plant material handling costs D. H. Driscoll (*Iron Steel Eng.*, 1959, **36,** Oct., 133–137) The use of a road truck system for materials handling at Colorado Fuel and Iron Corp. is described, and the reduction in delays and increased utilization of storage space attained are indicated.

A.C. versus D.C. for crane hoist drives A. Straughen (*Iron Steel Eng.*, 1960, **37,** Feb., 132–136) Cost, efficiency, and safety are considered.

10-Ton magnet cranes for a beam mill Wellman Smith Owen Eng. Corp. (*Engineer*, 1959, **208,** Dec. 11, 782–784) An illustrated account of the installation at the Lackenby mill. The cranes handle beams up to 36 in. ×

16½ in. × 260 lb/ft length, 16–90 ft long. Plans are shown.

Special 'Craven' ingot slicing machine (*Craven Machine Tool Gazz.*, 1960, Jan.–Feb., 4–5).

Equipment crops bar stock cutoff costs Ingersoll Milling Machine Co. (*Steel*, 1960, **146,** Feb. 15, 126) The use of an automatic cut-off saw for round and flat rolled shapes and forgings is described. The machine can be set to a programme, leaving the operator free for other work.—D.L.C.P.

Turnover device handles slabs Empire-Reeves Steel Corp. (*Iron Age*, 1960, **185,** Feb. 11, 153) A device of a bar supporting four endless chain slings used two at a time for slabs under or over 14 ft is described.

WELDING AND FLAME CUTTING

Automatic welding in the United Kingdom D. B. Tait (*Canad. Metalw.*, 1960, **23,** March, 36–38, 40) The continued covered electrode process is described with an outline of the submerged arc process.

Welding processes applicable to carbon and alloy steels H. Thielsch (*Canad. Metalw.*, 1960, **23,** April, 37) A table showing the processes commonly, occasionally, and not used for various grades of iron and steel.

The ABC's of stainless steel welding H. Thomasson (*Canad. Metalw.*, 1960, **23,** April, 35, 55) A list of the usual types is given with notes on properties and uses and a general account of the effect of welding on corrosion properties.

The welding of ductile iron G. R. Pease (*Weld. J.*, 1960, **39,** Jan., 1s–9s) The properties of S.G. cast iron (ductile iron) are reviewed, and the difficulties involved in welding this type of iron are discussed. The various types of welding, and the success obtained by their use, are considered. It is concluded that, at present, it is possible to obtain fairly satisfactory welds —but only with post-weld heat treatment— but that welds of 'base metal' strength have not yet been attained. Lines of investigation for the future are suggested.

Aspects of the welding of cast iron C. Pfeiffer and S. Lundin (*ESAB Svetsaren*, 1958, **23,** (3), 51–57) Investigations by the authors on the welding of nodular cast iron are reported and discussed.

Welding of ferritic annealed nodular cast iron W. Zitzelsberger (*Schweissen Schneiden*, 1959, **11,** Nov., 416–427) Welding of nodular cast iron presents the same difficulties as other types of cast iron, with the additional factor that the welded joint must in some cases possess the same deformability as the initial material. Various welding methods and materials are examined critically.

Repair welding of a heavy chilled iron roll E. Friedrichs (*Schweissen Schneiden*, 1959, **11,** Oct., 402–403) Details of the operation are given—the roll was of alloy cast iron, containing 1·5%Cr and 0·5%Mo.

Recommendations for the choice and classification of steels for use in welded structures (*Brit. Weld. J.*, 1959, **6,** Nov., 530–533) Recommendations are made for the choice and classification of steels for electric arc welding with covered electrodes of normal penetration. The steels included are rolled mild steels and Mn steels.—T.G.

Repair welding of steel castings S. Lundin (*ESAB Svetsaren*, 1959, **24,** (3), 39–45) Techniques of welding plain C and alloy cast steel are reviewed, and some examples given of the repair welding of heavy castings.

Welding of chromium steels H. Gerbeaux (*Soud. Techn. Connexes*, 1960, **14,** Jan.–Feb., 39–61) Advantages and applications of various Cr steels are discussed. The weldability of several kinds of Cr and of Cr–Mo steels is dealt with. It is concluded that welding of hardening and self-hardening Cr steels requires the use of basic electrodes and preheating as well as tempering. The welding of ferritic Cr steels is inadvisable in the case of heavy loads (15 refs).—R.F.

Automatic welding of 15Kh1M1F steel under carbon dioxide B. S. Kasatkin and Yu. N. Vakhnin (*Zavods. Lab.*, 1959, (11), 13–19) This heat-resistant pearlitic steel can be welded under CO_2 with special electrodes. The joints have satisfactory mechanical properties between 20 and 570°C. The long-term strength and fatigue limits of the joints are similar to those of the base metal. Between 570–620° the joints display a stable structure in long-term ageing.

Welding high strength steels with fused fluxes A. M. Makara, T. M. Slutskaya, and N. A. Mosendz (*Avtom. Svarka*, 1959, (10), 3–8 In welding these steels with fused low-Si and low-Mn AN–15 flux, high quality joints may be obtained with small O and P contents in the weld metal. The impact strength of these joints is not below that of joints with ceramic fluxes. Flux AN–15 has good technological properties and may be used for welding small pipes. Its composition is 23·8–28·8%SiO_2; 1·8–2·2%MnO; 14·4–18·4%CaO; 9·0–11·0% MgO; 21·7–24·9%Al_2O_3; not more than 0·85%FeO; 0·03%S; 0·03%P, and 20–22·5% CaF_2.

Fabricating stainless steel A.P.V. Co. Ltd (*Mach. Lloyd*, 1960, **32,** Jan. 30, 67–68) An illustrated account of welding and other processes.

The metal-arc welding of stainless steels: Part II M. C. T. Bystram (*Welder*, 1959, **28,** July–Sept., 56–62) With a table of the standard range of electrodes.

Notes on the welding of steels having protective coatings M. K. Williams (*Welder*, 1959, **28,** Oct.–Dec., 70–71) Clad, Al, and Zn coated and painted sheets are briefly reviewed.

Mechanization of bare arc welding with powder wire I. K. Pokhodnya and A. M. Suptel' (*Zavods. Lab.*, 1959, (11), 3–12) A method is described of mechanizing bare arc ac or dc welding with powder electrodes containing slag and gas-forming substances. The cost of 1 kg of weld metal is the same as that of consumable electrode welding under CO_2 and cheaper by a factor of 1·8 to 3 than that of hand electrode welding.

Argonarc welding J. Novotný (*Zváranie*, 1959, **8,** (9), 276–282) [In Slovak] Argonarc welding as practised in Czechoslovakia, with consumable and W electrodes, is discussed. Welding equipment developed in recent years is considered.—F.F.

Arc welding processes and equipment for automation F. E. Adamson (*Weld. J.*, 1959, **38,** Dec., 1070–1079) The various available methods of welding with automation are discussed and compared, and examples of their uses are given.

Steam as a shielding atmosphere in arc welding L. S. Sapiro (*Avtom. Svarka*, 1959, (10), 92–93) At the Ukraine LKSM works a process of arc welding has been evolved and is used for the arc welding of MSt 3 steel using 2 mm dia. welding wires type SV–08, and steam from a special boiler from which the volume can be strictly controlled. The 280 amp 35 V current is supplied by a standard generator. The rate of welding reaches 25 m/h with reverse polarity.

Butt welding steel sheet by the foil-seam process L. W. Mecklenborg (*Weld. J.*, 1960, **39,** Jan., 19–28) Work on the welding of mild steel sheet by this process is described. The quality of welds made by this process is excellent, and other advantages are strong, leakproof joints, minimum thickening, excellent appearance, ease of finishing, high welding speed, and minimum deformation of material. The effect of welding current, welding speed, gap between the edges of the butted workpieces and the gauge of the workpieces was investigated. They are interrelated, but, if the welding current range which produces welds showing 100% joint efficiency in tension is considered to be a measure of welding latitude, then, for a given level of the two remaining variables, the welds become more critical with increase in (1) stock thickness (2) speed, and (3) gap between the edges to be welded.

Experience in the welding of oil well apparatus made with two layered steels A. G. Fomin (*Avtom. Svarka*, 1959, (10), 85–87) The technique of welding units for oil well installations made of two layers of steel types St 3 and 12 MKh, or of stainless 108Kh12 is described and the conditions of operation and the specification of the electrodes are given.

Welded pinion stands offer economies over conventional cast units C. J. Schorsch (*Iron Steel Eng.*, 1960, **37,** Jan., 147–148, 150, 153) A description is given of a 28-in. welded pinion stand, one of four which form part of a hot strip mill being constructed for August Thyssen Hütte by United Engineering and Foundry Co., Pittsburgh.

The conditions for contact spot welding of low carbon steel machine components S. M. Tas'ba (*Zavods. Lab.*, 1959, (12), 28–33) An expression has been derived for the principal parameters of spot welding low carbon steels which, depending on the thickness of the work piece, makes it possible to calculate the dia. of the working surface of the electrode, $d_e = 5{\cdot}5\sqrt{\delta}$; the pressure on the electrode $P_e = 200\,\delta$; and the strength of the welding current or its time of passage $\frac{I^2 V}{10^6} = 22\delta^2 + 10$.

Production of welded tubes of large diameters in the German Federal Republic and France A. I. Tselikov and B. E. Paton (*Stal'*, 1960, (3), 243–253) Gas main production by cylindrical or spiral strip welding is reviewed.

Calculation of the length of the heating zone in the electric roll butt-welding of tubes B. D. Zhukovskii (*Lit. Proizv.*, 1959, (10), 29–31) The principal parameter in this operation for calculating the welding current and other factors is the time of heating of the metal, which for a given speed of the operation is determined by the length of the welding focus (the heating zone). The theoretical considerations for evolving a suitable formula are given and a comparison is made between calculated and actual heating zones for different sizes of tubes which agree within a reasonable approximation.

Welded tubing measures up for atomic reactor use Standard Tube Co. (*Steel*, 1959, **145,** Nov. 23, 90–91) Square-section tubing was made to tight specifications for use in atomic reactor fuel cores by careful choice of conventional techniques. The tubing was 2·6 in. square, 0·093 in. wall thickness, 6 ft lengths. Stainless steel strips were round formed, welded, rough shaped, drawn square, and straightened.—D.L.C.P.

High-frequency equipment speeds tube welding New Rochelle Tool Corp. (*Iron Steel Eng.*, 1959, **36,** Nov., 164, 166, 169) An electronic device can be used for the production of thin-walled welded tube from sheet in a variety of metals, at speeds greater than 250 ft/min. The principle of the method is described, and several users of this Thermatool process are enumerated.

Site welding of large-diameter studs at the Velindre cold mill R. W. Taylor and A. G. Senior (*Brit. Weld. J.*, 1959, **6,** Nov., 514–520) The welding of steel studs for the attachment of rails to elevated crane gantries is described including the fatigue tests carried out for the determination of the most suitable form of stud.—T.G.

Puddle welding British Oxygen Co. Ltd (*Aircraft Prodn.*, 1960, **22,** April, 145–148) An account of the argon-arc fusion spot-welding process for very thin stainless steel for flutter models.

A new electro-slag welding process D. J. W. Boag and W. K. B. Marshall (*Brit. Weld. J.*, 1959, **6,** Nov., 507–513) The Rockweld–Vus 'Vertomatic' electro-slag welding machine, a Czechoslovakian development of the originally Russian welding process, is described. Operation and application, effect of welding variables, speed of welding, and the properties of the welds and comprehensively reviewed.

Method of welding high-alloy steels and alloys under flux Ya. Lukashek and K. Lebl' (*Zavods. Lab.*, 1959, (12), 3–12) The method described provides for feeding the wire electrodes arranged parallel to the work piece. The proportion of base metal in the weld metal has been reduced and productivity increased. This method may be used for welding stellites to the surfaces of fittings for service in steam under high pressures.

Electro-slag welding of ageing chromium–nickel–aluminium stainless steels B. I. Medovar, A. N. Safonnikov, E. Ya. Belkin, and O. A. Sharov (*Avtom. Svarka*, 1959, (9), 33–44) When welding CN–2 steel with an electrode of the same metal of large cross-section, the appear-

ance of non-welded places is probable. Any fusion of the weld edge practically attainable would not eliminate this fault where a two-component fluoride flux with a large proportion of Al_2O_3 is used. To overcome this defect, fluoride fluxes containing much alumina must not be used.

Electro-slag welding of heat-resistant chromium steels with lamellar electrodes B. I. Medovar and A. N. Safonnikov (*Zavods. Lab.*, 1959, (12), 80–82) The Paton Institute has worked out conditions for the electro-slag welding of two types of heat-resistant steels (EI 961 and EI 736) in various thicknesses and the suitable electrodes, current strength, rate of feed, and voltages are tabulated.

Pressure welded austenitic–ferritic composite junction pieces (Mannesmann-type joints) H. Linden and H. Hennecke (**Mitteilungen der VGB*, 1958, (54), June, 165–175) Extensive tests on welds for high pressure boiler plant are reported. The method of production and the effects of temp. changes and temp.-time effects on C diffusion and the development of routine tests are described. Metallographic observations, heat treatment, and stress corrosion tests are also discussed.

Power for the friction welding of steel rods V. I. Vill' (*Svar. Proizv.*, 1959, (10), 12–15) The third stage in friction welding is conditioned by the amount of heat evolved. To increase the efficiency of friction welding it is not necessary to increase the speed of rotation. With small rotation speeds the duration of the initial period of heating increases considerably and therefore also the condition of the surfaces for welding change. For any given condition there corresponds an optimum speed of rotation. Given a maximum frictional force the motor required is approximately 1·5 times more powerful than for normal welding processes. It is therefore necessary so to alter the axial force that maximum friction can be reached with relatively low specific pressures, as, for example, by gradually increasing the force. This suggestion is based on experiments with the friction welding of 8–50 mm diameter steel rods at a speed of 240–3000 rpm with a specific pressure of 2–10 kg/mm².

Heating and cooling of rods butt-welded by friction N. N. Rykalin, A. I. Pugin, and V. A. Vasil'eva (*Svar. Proizv.*, 1959, (10), 15–18) These investigations aimed at elucidating the heating of the regions adjoining the contact surface and the initial distribution of the specific heat flow in the contact zone surface in friction butt welding 20 mm diameter rods of low-carbon steel with a specific pressure of 0·8–0·9 kg/mm² at 1400 rpm. A method has been evolved for calculating the thermal cycle necessary for the welding joint, and also the parameters of the operation for heating by means of a flat evenly distributed source of heat, not limited along the length of the rods, by side surfaces which impede the heat flow. These calculations come sufficiently close to the actual temp. measured during these experiments.

The power and heat required for the friction welding of thin-walled tubes A. S. Gel'man and M. P. Sander (*Svar. Proizv.*, 1959, (10), 18–20) With increasing temp. the friction coefficient increases rapidly in the first stages and then decreases gradually. An increase in the linear speed and in the specific pressure brings about a decrease in the friction coefficient and the power required changes therefore correspondingly. It is advisable to begin operation at a relatively small pressure (1 kg/mm²) and then when the temp. exceeds the value corresponding to the power peak, increase the pressure (by a factor of 5–6 for welding steel).

Friction welding as a method of repairing worn components Kh. A. Tyayar (*Svar. Proizv.*, 1959, (10), 23–24) It is possible to build up worn metallic surfaces of machine components by friction welding. For good results, the specific pressure must be adjusted to the calculated rotational speed of rod and component. An increase in speed for the sake of increasing productivity is not advisable since it results in uneven work. At great speeds the work-piece has no time to heat up and the built-on layer separates from the cold metal and no welding occurs. The building up operation can be easily automated.

High vacuum-electron-beam welding E. B. Bas and G. Cremosnik (*Vacuum-Technik*, 1959, **8**, Oct., 181–188) Optical theory of the method is described in detail, and examples are given of welded joints made in stainless steel and some non-ferrous metals (10 refs).

Welded and expanded joints for high pressure pre-heaters H. Haferkamp (**Mitteilungen, VGB*, 1959, (6), Oct., 365–367) Steam pre-heaters are fabricated by expanding and welding the tubes into the tube plate. The process is described for plant with 60 atm. steam and 200 atm water. Various experimental methods of fabrication were tried and a satisfactory process using gas welding was elaborated.

Frame for welded marine engine G. Hellström (*ESAB Svetsaren*, 1958, **23**, (3), 37–50).

New fluxes of the Czechoslovak Welding Research Institute Š. Horváth and E. Pikna (*Zváranie*, 1959, **8**, (9), 261–267) [In Slovak] The compositions and properties of a number of fluxes recently developed for use in automatic welding of constructional and 18/8-type Cr–Ni steels are described.—P.F.

New laboratories of the Welding Institute at Saint-Ouen (*Soud. Techn. Connexes*, 1960, **14**, Jan.–Feb., 3–4) The main object of the new laboratories is the non-destructive testing of welded and other parts. The equipment includes several types of ultrasonic and X-ray installations, gamma radiography, and magnetic methods of testing.—R.P.

Welding properties of steels J. Lomas (*Mach. Lloyd*, 1960, **32**, April 16, 30–33) A brief review of all types.

Chemical-metallurgical studies on welds K. Born (*Schweissen Schneiden*, 1960, **12**, April, 139–146) The author discusses the usefulness of the modern isolation techniques in the inspection of welds where the residues allow deductions to be made as to the mechanism of the formation of endogenous inclusions. The cause of weld defects can often be deduced with a surprisingly high certainty. This test is of particular interest for the detection of sulphur in the weldment as this very often escapes detection when conventional methods are used owing to the high dispersion of the particles which have no time to coagulate during cooling of the weldment.—T.G.

The structure of low-carbon welded joints B. S. Kasatkin and G. F. Darovskii (*Avtom. Svarka*, 1959, (9), 13–16) As a result of tests here explained, it can be assumed that the macro substructure of low-carbon weld metal favours the inhibition of shear deformation and the development of microcracks. This must reflect favourably on an increase in tensile strength and plasticity and in a reduction of the tendency to cold shortness of the weld metal. This effect increases with the refinement of the substructure.

On the phase structure of weld joints of austenitic and ferritic steels I. Hrivňák (*Zváranie*, 1959, **8**, (9), 258–261) [In Slovak] Problems connected with welding of 17/7 Mn–Cr steels with ferritic electrodes were studied. It was found that a hard zone of bainitic structures forms in the transition between the ferritic and austenitic components of the weld, if interdiffusion is not inhibited. The morphology of the carbides and other components present in the transition zone was investigated. It is questionable whether joints of this type would be acceptable in practice, particularly at elevated temp., unless interdiffusion during welding is prevented.—P.F.

Stresses at the joint face of austenite–ferrite welds A. Schiller (**Mitteilungen der Vereinigun der Grosskesselbesitzer*, 1958, June, 176–180) Formulas are derived for the calculation of internal stresses at the face of butt welded or diagonally welded (Mannesmann type) tube joints.

Sigma phase in stainless welded material, type 19 Cr–9 Ni I. Wachtmeister (*ESAB Svetsaren*, 1959, **24**, (2), 17–28) Mechanism of sigma-phase formation is outlined particularly in welding; and the influence of the phase on tensile strength, notch-impact strength, and microhardness is discussed.

Effect of layer peening on the quality of austenitic welds F. G. Gonserovskii (*Avtom. Svarka*, 1959, (9), 81–87) Layer peening of austenitic welded joints cannot be considered satisfactory because in non-rigid joints it reduces the residual stresses insignificantly and in rigid joints it increases them, and also because it changes the mechanical properties of the weld metal by increasing tensile strength but reducing plasticity and impact strength. Peening pearlite joints sharply increases welding stresses.

An investigation of welded crack arresters R. J. Mosborg (*Weld. J.*, 1960, **39**, Jan., 40s–48s) Laboratory tests were carried out in order to investigate the possibility of arresting cracks in welded components by directing them into a tough strake of steel. The investigation was carried out for temp. ranging from −25 to +40°F. In one case the results of these tests are compared with results from drop-weight, explosion-bulge, and Charpy V-notch tests.

The effect of inherent stresses and plastic deformation on the properties of the weld metal L. I. Gladshtein (*Lit. Proizv.*, 1959, (10), 27–29) The structure transformations which are connected with a change in the rate of cooling are not the only cause of changes in the mechanical properties of the weld metal on carbon steels. Welding stresses and the ensuing plastic deformation exert their own influence. Under their action the yield-point of the weld metal increases considerably (14–19%) and on the tensile stress curve there appears a plastic section which shows that ageing is taking place in the weld metal.

Continuous cooling transformation diagrams for welding of Mn Si type 2H steels H. Sekiguchi and M. Inagaki (*Tetsu to Hagane*, 1959, **45**, Sept., 1106–1108).—K.E.J.

Desulphurizing capacity of fluxes in electro-slag welding Yu. M. Zaitsev and G. S. Tyagun-Belous (*Avtom. Svarka*, 1958, **11**, (11), 57–60) Tests on various fluxes were carried out in the remelting of steels and in the welding of steels and cast iron. Welding current has an effect as well as slag composition.

Hydrogen in welding P. Bastien (*Lastechniek*, 1960, **26**, March, 56–65) [In French] This is a report from the International Institute of Welding in which the author discusses at length the relative behaviour of welded, cast, and forged steels. Reference is made to the causes leading to the appearance of H_2 in the steel and the resultant effects exerted on the physical properties of the latter. To conclude, the possible mechanism of the brittleness caused in steel by the presence of H_2 and its consequences when carrying out welding operations are also enlarged upon.—F.R.H.

On the prevention of cracking in the welding of fully austenitic stainless steels F. Tsukamoto, H. Sunaga, and K. Nagayoshi (*Tetsu to Hagane* 1959, **45**, Sept., 1010–1013).—K.E.J.

Evaluation of weld-joint flaws as initiating points of brittle fracture. Part II R. P. Sopher, A. L. Lowe jun., D. C. Martin, and P. J. Rieppel (*Weld. J.*, 1959, **38**, Nov., 441s–450s) This work was carried out to determine the conditions under which flaws of less than 4 in. in length would initiate brittle fractures. It appears that flaws between 2 and 4 in. may initiate brittle fractures in a sufficiently triaxial stress field, while 4 in. flaws at least are required in a biaxial stress field. The very important part played by residual or reaction stresses was stressed.

Origin of weld cracks in a railway bridge A. Fabiszewski (*Przeglad Spawalnictwa*, 1958, **7**, 199–202; from *Polish Techn. Abs.*, 1959, (3), 31) The cracks were detected on X-ray inspection and were caused by wrong welding technique.

Detection of damage due to excess oxygen in gas welding H. Fiehn (*Schweissen Schneiden*, 1959, **11**, Oct., 393–396) The use of notch tensile specimens in assessing this type of damage is recommended in preference to other methods, as excess O_2 in the welding flame causes a marked loss of deformability.

Joining and fastening of materials (*Met. Design. Eng. Materials Selector Issue*, 1959–60, 1959, **50**, 431–453) Methods are tabulated for ferrous, non-ferrous, and non-metallic materials (432–433), brazing and soldering alloys are set out (434–437), also welding rods (438–445).

Brazing stainless steel to copper M. Liebson (*Rev. Sci. Instr.*, 1959, **30**, May, 369–370) Technique is described and solders are recommended for the fabrication of electron tubes.

Thermal fatigue and high-temperature brazements R. L. Peaslee (*Weld. J.*, 1960, **39**, Jan., 29–33) Where a thin metal part is subjected to thermal fluctuations and at the same time it is constrained by thicker stiffeners or supports, it has been found that the thin part may fail rapidly due to the stresses involved. Some cases in which this type of failure has arisen are described and the type of design required to avoid this sort of failure is explained. It is pointed out that, as the service temp. of brazed parts reach 1000–1500°F, this type of failure may grow more prevalent, and it should be investigated more thoroughly before this happens.

High-frequency induction brazing and soldering J. Libsch and P. Capolongo (*Weld. J.*, 1959, **38**, Nov., 1059–1069) The properties of brazing and soldering alloys and fluxes are considered, and the advantages of the high-frequency method are pointed out. The design and equipment required for H-F methods are described.

Graphite cloth speeds heat cycle Rohr Aircraft Corp. (*Steel*, 1960, **146**, Feb. 22, 74) The time for brazing stainless steel honeycomb panels has been cut from hours to minutes by employing graphite cloth resistors instead of large furnaces. The cloth is laid over the assembled panel in a brazing chamber, copper bus bars are hooked on to the ends of the cloth, the chamber is flushed with argon, and power is turned on until brazing temp. is reached.

Adhesive bonds and their susceptibility to corrosion O. Blahož and M. Jakob (*Sbornik (Ostrava)*, 1959, (6), 673–681) The properties of the resin 'Epoxy 1200' were found to render it particularly suitable for the preparation of bonded metal-to-metal joints, in cases where no alternating or bending loads were applied. The corrosion susceptibility of the joints was studied, and corrosion was found to affect the strength of the joint significantly. Corrosion protection of bonded joints is recommended, and methods of protection are indicated.—P.F.

The fastening of cemented carbides O. Knotek (*Werkstattstech.*, 1960, **50**, April, 173–176) Soldering—being still the most important and most widely used method—clamping, shrink-fitting, bonding, and welding of cemented carbides are briefly reviewed. In a table are listed the most common solders and their melting-points.—T.G.

Self-centring tip for oxy-acetylene cutters B. I. Smirnov (*Lit. Proizv.*, 1959, (10), 39–41) A description is given of a new appliance, a tip for the oxyacetylene cutters, easy to mount and ensuring a concentrated flame for which various advantages are claimed.

The cutting of metals and alloys other than plain carbon steels G. Gronier (*Techn. Mod.*, 1959, **51**, Dec., 598–601) Ferrous alloys to which reference is made are stainless and other alloy steels. The technique of oxygen cutting using iron powder (The 'Safjet' process) is described, together with some applications.

Powder lancing in steel mill operations C. B. Milton and L. F. Keel (*Iron Steel Eng.*, 1960, **37**, Feb., 95–101) The equipment and its use for cutting and piercing operations in various iron and steelworks operations and maintenance work are described.

Effect of the shape of the nozzle on the cutting speed in flame cutting A. Matting and K. R. Choudhury (*VDIZ*, 1960, **102**, April 21, 459–466) Based on a study of eight commercially available nozzles the authors developed a special nozzle for square cutting. This nozzle gives a very high cutting speed by using O_2 at a rather high pressure but the quality of the cut face is not impaired by the high rate of cutting. The undesired formation of martensite at the cut face has been avoided by giving the nozzle a special shape. A new nozzle was also developed for oblique cutting.—T.G.

A comparison between oxy-acetylene and oxy-arc gouging H. von Hofe and S. Piwowar (*Schweissen Schneiden*, 1960, **12**, April, 146–153) The studies made for arriving at an appraisal of the methods are described in detail. The results may be summed up as follows: The oxy-acetylene process is superior in all respects to the arc processes with or without the use of O_2, it cannot, however, be applied to high-alloy and stainless steels and cast iron, but the application of the arc processes is also restricted in these cases. The main points are that the torch is easier to handle, gives a better and more consistent cut, the surface quality is better, and mechanization is easier to achieve. The costs are also lower.—T.G.

MACHINING AND MACHINABILITY

Chart of comparable tool steel brand names Uddeholm Co. Inc. (*Pamphlet issued by Uddeholm Co. of America Inc.*, pp.2 N.D.) SAE numbers and brand names of 13 steel companies are given.

How to evaluate cutting tools C. J. Oxford jun. (*Metalw. Prodn.*, 1960, **104**, April 6, 610–612) Statistical considerations in the interpretation of tool life test results are briefly considered.

A new factory for milling cutters of super high-speed steel Larcher Cutters Ltd (*Mach. Lloyd*, 1960, **32**, April 16, 48–49) Notes on high-speed steels are given.

Twenty tools slice cast steel ingots Taylor Bros. & Co. Ltd. (*Metalw. Prodn.*, 1960, **104**, April 6, 621) A machine for cutting blanks for railway wheel tyres is shown.

Sheet metal data sheet: Shearing sheet metal J. W. Laugton (*Sheet Metal Ind.*, 1960, **37**, Jan., 63–65) Typical machines and their operation are briefly indicated.

Separating the roughing operations from turning on the lathe R. Radulescu (*Met. Constr. Masini*, 1959, (11), 983–984) By this separation it is claimed that greater efficiency was secured together with a substantial saving of metal and costs.

Model experiments with machine tool parts J. Bielefeld (*Schweissen Schneiden*, 1959, **11**, Oct., 396–401) A detailed investigation of factors influencing the rigidity of structural elements, including the effects of stiffeners.

Hydraulically-operated bar skimming machine (*Craven Machine Tool Gazz.*, 1960, Jan.–Feb., 1–2, 8).

Our chip making inefficiency is a national disgrace H. Frommelt (*Steel*, 1959, **145**, Nov. 23, 78–80) The author explains how the widespread aim of obtaining a long cutting-tool life results in unduly high machining costs. Often the cost of reduced tool life resulting from higher cutting speeds would be more than offset by a bigger return for plant and labour expenses.—D.L.C.F.

Influence of the structure on the machinability of structural steels W. Knorr (*Hart. Techn. Mitt.*, 1959, **13**, (3), 201–226) From the investigations described it is concluded that mild and case-hardening steels machine best with a well-differentiated ferrite-pearlite structure. Conditions are established for the optimum machinability of various alloy case-hardening and other alloy steels (13 refs).

A study on leaded free-cutting steels. III. On the properties of some leaded free-cutting steels C. Asada, T. Fujiwara, R. Kadowaki, and T. Muraji (*Tetsu to Hagane*, 1959, **45**, Sept., 1032–1034). **IV. Cutting mechanism of leaded free-cutting steels. I.** Y. Honda and T. Ito (1035–1036). **XI. Consideration on the effective lead contents in leaded special steels** T. Araki, A. Koyamagi, and H. Ohashi (1036–1038). III. The effects of total Pb on the properties of five steels are given, and the effects of particle size and shape of Pb are studied. IV. Data on cutting forces, friction, shear stress, etc., are given and compared with those for non-leaded steels. XI. Inclusions and notch factor distribution are correlated with Pb content.—K.E.J.

Spark erosion machining O. Rüdiger and A. Winkelmann (*Techn. Mitt. Krupp*, 1959, **17**, (3), 156–158) Design and operation of equipment are reviewed, and characteristic data for various electrode and workpiece materials at different machine outputs are enumerated.

Experience with spark erosion in finishing processes E. Wittwer (*Werkstattstechn.*, 1959, **49**, Oct., 613–618) Equipment and techniques are described with examples.

CLEANING AND PICKLING

Abrasives in metallurgy (*Aciers Fins Spéc.*, 1960, March, 46–50) The nature of coated abrasives is discussed and their applications in industry outlined. Examples quoted include abrasive belts operating on a free loop and those supported by a contact wheel, grinding, cleaning, and polishing of various metal parts.

New developments in blast cleaning G. W. Powell (*Foundry*, 1960, **88**, Jan., 84–87) A fully automatic blast cleaning machine is described which has been improved durin the last two years to such an extent that its use is now an economic proposition for the small foundry.—T.G.

Review of the process of dressing and shot peening using steel shot L. Gascuel (*PACT*, 1959, **13**, Dec., 500–501) Mechanism and equipment are briefly described.

The not-so-ideal cleaning cycle S. Spring (*Met. Fin.*, 1959, **57**, Oct., 64–68) Measures are described by which time can be saved and the number of processing steps reduced in cleaning work for plating.

How to get the most from solvent-vapour degreasing—part I T. J. Kearney and C. E. Kircher (*Met. Prog.*, 1960, **77**, April, 87–92) This section describes equipment designed to use sprays, warm solvent baths, and ultrasonics, and gives some applications. Brief reference is made to the incorporation of solvent degreasing and flow coating in a single unit.

Chemical cleaning of equipment in refineries and petroleum plants L. N. Klinge and J. Selman (*Corrosion*, 1960, **16**, Jan., 9t–18t) The authors discuss the advantages of chemical cleaning over mechanical methods in refineries and petrochemical plants. The cleaning of columns, coolers, condensers, heat exchangers, boilers, and drums is described in detail, including the potential corrosion difficulties encountered (258 refs).—T.G.

Cleaning and preparation of metals prior to electroplating. XII. Effect of fatty acid films H. B. Linford and J. J. Grunwald (*Plating*, 1959, **46**, Sept., 1039–1045) The effect of fatty acid films on the adhesion of Ni deposits on Cu has been quantitatively investigated. It is concluded that the presence of even small traces of organic contaminants on surfaces to be Ni plated will result in pitting due to H_2 bubble attachment. The presence of such contaminants on the cathode will result in preferential discharge of H_2 during the first fractions of a second of plating. The experiments are fully reported and discussed.—T.G.

Drum-type pickling of coiled strip L. W. Passmore (*Sheet Met. Ind.*, 1960, **37**, Jan., 66–68) A drum in which the coil is partly unrolled and re-rolled on reversal of the direction of rotation is described.

Smut residue measured after pickling of steel (*Ind. Fin.*, 1959, **11**, Oct., 32) A new type of ultrasonic cleanliness tester, designated CM–1, has been developed by Graham Research Laboratories of Jones Laughlin Steel Corp. It enables numerical values to be assigned to surface cleanliness, and it can be used to measure smut residue after pickling of steel, to compare different cleaning techniques and equipment, and to determine the working life of detergents and solvents.—G.F.

The effect of pretreatment on the scaling of metals. VI. The scaling behaviour of mild steel sheet after pickling in various oxy-acids A. Petzold (*Metallrein Vorbeh.*, 1960, **9**, April, 53–56) Scaling behaviour of mild steel at 800–900°C is not much affected by previous pickling in oxy-acids such as dilute or concentrated H_2SO_4, H_3PO_4, CrO_3, etc. Previously published reports have indicated a slight improvement in the scaling behaviour after lengthy pickling in dilute HNO_3.—T.G.

Will chemical displace mechanical polishing? (*Metalw. Prod.*, 1959, **103**, Oct. 16, 1639–1644) Some applications, mainly Continental, and non-ferrous, are given of chemical polishing, and reference is made to the work done by Laporte Chemicals Ltd. to develop on a commercial scale the use of Marshall's solution for polishing low-carbon steels.

The effect of the composition and structure of carbon steels on electro-polishing S. M. Tiranskaya (*Ukrainsk. Khim. Zhur.*, 1959, **24**, (4), 533–540) By increasing the anode potential, conditions are created for the change of the metal from the active to the passive state owing to the formation of oxide films of different compositions, during the first jump in potential with bivalent iron, and during the

second intermittent change with tervalent iron. The best of the three electrolytes studied is one of 1·78 sp.gr. containing 1000 ml of H_3PO_4 of 1·62 sp.gr. and 200 g CrO_3.

Ultrasonics conquers corrosion T. Fetisov (*Izvestiya USSR*, 1957, (275), translation from *US Dept. Commerce Techn. Trans.*, 1959, **2,** Oct. 2, 527) Rust removal by ultrasonics appears also to confer temporary anticorrosive properties.

Barrel finishing—new indications of its value (*Ind. Fin.*, 1959, **11,** Oct., 33–34) Several examples of the successful use of barrel finishing in industry are described, showing how the process can reduce costs while improving surface finish and fatigue strength.—G.F.

Frosty finish for stainless steel (*Ind. Fin.*, 1959, **11,** Oct., 32) A new frosty surface finish ('Fro-Zon') for stainless steel has been developed by Stamping Service Inc. of Detroit. It is made possible by a semi-blasting technique producing small mounds on the surface.

An electron diffraction study of electropolished iron P. B. Sewell, C. D. Stockbridge, and M. Cohen (*Can. J. Chem.*, 1959, **37,** Nov., 1813–1819) Electronic diffraction patterns obtained from Fe electropolished in an acetic–perchloric bath showed an 'atomically smooth' surface under optimum conditions, and evidence is shown for the existence of a thin surface film which thickens on exposure to air, forming a preferentially oriented cubic iron oxide.

The sulphuric acid recovery plant in the acid circuit of a wire pickling plant W. Mallach (*Wire*, 1959, (42), Aug., 102–109) The method is described of incorporating an $FeSO_4.7H_2O$ recovery plant in a large works with H_2SO_4 pickling plants in various locations, the effects due to salts and acid concentrations in the different pickling tanks being taken into account. Economic aspects are considered.

The treatment of trade effluent at Wilmot-Breeden Ltd., Bridgewater K. G. Robinson (*Metallurgia*, 1960, **61,** Jan., 3–5) The plant treats phosphating, anodizing (CrO_3–H_2SO_4), metal pickling, Cr-plating, and occasional acid effluent from clearing the vats, alkaline degreasing swills, cyanide residues, and occasional alkali and cyanide effluents from work vat dumping. It is situated in a rural area with stringent limits on effluent discharge.

PROTECTIVE COATINGS

Finishes and coatings (*Met. Design Eng., Materials Selector Issue*, 1959–60, 1959, **50,** 305–325) Tabulated properties of electrodeposited, sprayed; hot-dip, diffused, vapour deposited, enamel, and organic coatings are set out.

'Close-plating'—a page from the past J. W. Tomlinson (*Met. Fin.*, 1959, **57,** Oct., 62–63) An illustrated description of the process.

The effect of protective coatings on the properties of ferritic steel at 650° C K. W. Mitchell, H. King, and E. A. Brandes (*Metallurgia*, 1960, **61,** Jan., 15–23) Tests on mild and Mo–B, Mo–V, and Cr–Mo–V–W low alloy steels are described. Al coatings were applied by hot dipping or spraying followed by diffusion treatment and Cr by deposition from chloride. Only Mo–V steel showed no fall in creep and fatigue strength but oxidation resistance was improved. Chromizing produced a loss of core strength which could be partly restored by heat treatment.

The influence of electrolytically deposited protective coatings on the fatigue strength of steel H. Tauscher (*Draht*, 1959, **10,** Oct., 511–516) After a general introduction, the influence of H_2 introduced during plating, in static and alternating stressing is discussed, and the effects of hard Cr, Zn, and Cd plating, and other coatings, on fatigue strength are examined with reference to the literature. A final section deals with the simultaneous effects of alternating stresses and corrosion.

The function of the conveyor in the finishing shop M. Reeves (*Electropl. Met. Fin.*, 1960, **13,** Jan., 10–14) The use of conveyors in electroplating and paint finishing, and the continuity, uniformity, and low reject rates made possible by their use are discussed.

Evaluation of adhesion of thick electrodeposits H. Chessin and J. G. Poor (*Plating*, 1959, **46,** Sept., 1037–1038) A destructive method has been developed for determining the quality of adhesion of Cr and other deposits to the basis metal. This test is also applicable to thick production pieces that cannot easily be subjected to bending tests or to brittle basis metals such as cast iron. The mechanics of the test are simple and fast.—T.G.

Filtration of electroplating solutions R. W. Gray jun. (*Plating*, 1959, **46,** Oct., 1165–1170) Types of filter are described and factors in choice of method are discussed.

The removal of carbonates from cyanide baths by means of ion exchangers A. Knodler and F. Elser (*Metalloberflache*, 1959, **13,** Oct., 327–330) It is shown that good results are obtainable by cation exchange only with low $(CN)^-$ contents, high $(CO_3)^{--}$ contents, and relatively low pH values. Strongly alkaline baths are unsuitable, and in all baths some loss of metal ions is to be expected.

Investigations of thin electrodeposited layers E. Raub, M. Mietka, and H. Beeskow (*Metalloberflache*, 1959, **13,** Oct., 321–327) Layers of Cr, Ni, and Ag 50–350 Å in thickness on highly-polished, polycrystalline Cu sheet were examined by diffraction and electron microscopy.

The nature and effect of the porosity in electrodeposits. V. An evaluation of the sensitivity of the ferroxyl test F. Ogburn, D. W. Ernst, and W. H. Roberts (*Plating*, 1959, **46,** Sept., 1052–1053) Strausser's modification of the ferroxyl test was used to determine the porosity of gold coatings on steel. It was found that less than 50% of all pores present were detected by the ferroxyl test. The larger the pore the greater the chance of its detection by this method.—T.G.

Restoring confidence in chromium plating D. J. Fishlock (*Metalw. Prodn.*, 1960, **104,** April 6, 601–607) Both Cr and Ni are discussed in regard to recent unpopularity of plated finishes. It is concluded that good plating practices are readily available and quality should improve even beyond the best present standards.

Bright nickel plating Mond Nickel Co. (*Bibliography No.*59/6, 1959, July, pp.54).

Decorative high-chloride nickel plating M. Tippman (*Plating*, 1959, **46,** Sept., 1054–1056) The author reports experience gained over many years with high-Cl Ni plating baths and describes the process in detail. Corrosion resistance and appearance compared with any bright Ni from other solutions for decorative work.—T.G.

Automatic plating installation for the processing of screws and other small parts W. Peters (*Draht*, 1959, **10,** Sept., 481–484) The type of plant described as 'Universal Automat' is described, in which different groups of products can receive modified treatments, the groups being separated by the transport system which charges the bath cells. Examples are given of equipment and programmes for Ni and Sn plating.

The anodic behaviour of various grades of electrolytic nickel E. Raub and A. Disam (*Metalloberflache*, 1959, **13,** Oct., 308–314) Nineteen grades of electrolytic Ni were investigated, and anode potential–current density curves are reproduced for several electrolytes.

Effect of nickel plating on the fatigue strength of steel Mond Nickel Co. (*Bibliography No.*59/2, 1959, July, pp.7).

The electrodeposition of nickel alloys (excluding proprietary solutions such as nickel-cobalt, nickel-phosphorus, nickel-tin) Mond Nickel Co. (*Bibliography No.*59/3, 1959, July, pp.20).

Stress in electrodeposited nickel Mond Nickel Co. (*Bibliography No.*59/4, 1959, July, pp.14).

Electroless plating techniques Mond Nickel Co. (*Bibliography No.*59/5, 1959, July, pp.27).

An outline of the chemistry involved in the process of catalytic nickel deposition from aqueous solution. Part I G. Gutzeit (*Plating*, 1959, **46,** Oct., 1158–1164) The reactions taking place in the electroless deposition process with H_2PO_2 ion are set out and the kinetics are evaluated.

Symposium on electroless nickel plating. History of the electroless plating process A. Brenner (*ASTM STP. No.*265, 1959, 1–2) The early studies on the reduction of Ni and Co ions to metallic Ni and Co by hypophosphites are described, and the way in which this led to the production of metallic 'mirrors' of Ni and finally to the electroless plating of Ni is outlined. **Chemical reactions** G. Gutzeit (3–12) The fundamental reactions, reaction kinetics, life of coating baths, bath stability, and bath composition are discussed. **Characteristics of deposits** W. H. Metzger jun. (13–20) The analysis, appearance, structure, physical, mechanical, electrical, and magnetic properties, resistance to corrosion and various reagents are discussed and tabulated. **Processing procedures** A. Krieg (21–37) The compositions of acid and alkaline baths, the quality of the chemicals used, the avoidance of spontaneous decomposition, brighteners, bath control, operating conditions, batch, barrel and continuous methods of operation, the possible basis materials and pre- and post-treatment processes are considered. The Lustralloy, Kanigen, and Alcoplate processes are discussed. **Advantages and limitations** E. B. Saubestre (38–40) The electroless Ni process is compared with the electrolytic plating process under the headings of solution make-up and maintenance, equipment required, plating characteristics (i.e. deposit uniformity, lack of porosity, need of catalysts, and slow deposition rate) and the properties of the deposit. **Applications** W. H. Safranek (41–49) This account is based on a survey of the literature, and on a questionnaire sent to about 20 companies using electroless nickel plating methods. Uses considered include corrosion protection, resistance to wear, abrasion, friction, and galling, improvement of fatigue properties of steels, reduction of stress-corrosion in stainless steels, joining of dissimilar Cr and Al alloys, facilitating the joining of ceramics to metals and attaching Ag and other electro-plates to Ti. It can also be used to build up worn and mis-machined parts. The cost of the process is considered in relation to that of electroplated nickel. **Test methods for electroless nickel coating and plating solutions** B. Cohen (50–52) The various tests for ductility, adhesion, hardness, corrosion resistance, and thickness are described. The test methods for the plating solution are also considered, and the procedures for analysis of the plating and the plating solution are given in detail in an appendix. **The patent situation** G. Gutzeit (53–62) A list of US Patents, complete up to August 1959, is given, with summaries of the principal allowed claims in each case. **Bibliography of references on electroless nickel plating** C. F. Waite (63–67) 136 references are given. Unlike the patent section, work in countries other than the US is referred to.

Cathode polarization during the formation of iron cobalt alloy and the causes of depolarization and over-voltage A. L. Rotinyan and E. N. Molotkova (*Zhurn. Prikl. Khim.*, 1959, **32,** (11), 2502–2507) In studying the cathode polarization during the formation, between 25–85°, of Fe–Co alloy from electrolytes of different concentrations of components settling out at the electrode at a pH of 3·5, a comparison of the partial polarization curves with those for the elimination of pure metals has shown that the deposition of the metal in the alloy is accompanied by a considerable shift in potential. The depolarization of iron and the overpolarization of cobalt depend on the composition of the cathode precipitates.

Galvanizing of cast iron (*Fonderie*, 1960, Jan., 28–29) Some difficulties encountered in galvanizing are pointed out and suggestions made in order to overcome these, particularly during pre-treatment.—R.F.

Zinc alloy for coating iron components by hot zinc plating N. N. Gratsianskii and P. F. Kalyuzhnaya (*Ukrainsk. Khim. Zhurn.*, 1959, **25,** (3), 382–384) A new composition is proposed for the bath for hot zinc plating, with an addition of Al–Si alloy containing 11–13% Si. It has been found that an addition of this alloy to the bath reduces zinc consumption per m^2 of surface from 370 to 300 g. A flux is also suggested for zinc plating consisting of 91% NH_4Cl, 4% NaCl, 2·5% NaF, and 2·5% glycerine.

Hot galvanizing of continuous strip by the

Sendzimir method W. Hettich (*Europaischer Mash. Markt.*, 1960, **10**, (2), 15–18) The process is outlined and the quality and uses of the strip produced are given.

A study of spangle in hot-dip galvanized coatings. I. Crystallization of hot-dip coated zinc melt on the iron sheet G. Mima, K. Kajikawa, and T. Kato (*Tetsu to Hagane*, 1959, **45**, Sept., 991–992) The nucleation and growth of spangles are studied photographically, and linear growth of no. of nuclei are examined as functions of time.—K.E.J.

Bright annealing wire galvanizing H. Banneitz (*Draht*, 1959, **10**, Oct., 522–523) A brief description is given of the operation and control system of an electric bright annealing furnace and of the subsequent galvanizing of the wire.

Comparison of methods for measuring the thickness of tin coatings on steel, with special reference to the use of isotopes J. Teindl and B. Otta (*Sbornik (Ostrava)*, 1959, **5**, (5), 533–547) Detailed studies of various methods used for checking and controlling the thickness of tin layers on steel were carried out, with a view, mainly, to ascertaining the most suitable process for use in practice. The isotope method is considered best for laboratory thickness determinations; for works practice, weighing of the sheet before and after tinning is considered most appropriate.—P.F.

Reaction of iron carbon alloys with molten tin H. H. Stadelmaier, L. J. Hütter, and W. K. Hardy (*Z. Metallk.*, 1960, **51**, April, 210–211) The ternary carbide Fe_3SnC_x was found in the alloy layer that is formed when Fe–C alloys such as cast iron react with molten tin at 350, 550, and 950°C. At 350°C the binary intermetallic compound is formed in the alloy layer.—T.G.

Improvements in metal finishing techniques Pyrene Co. Ltd (*Mach. Lloyd*, 1960, **32**, April 16, 27–29) Modifications of Parkerizing and Bonderizing are mentioned.

Structure and wear resistance of electrodeposited ferrous coatings Yu. N. Petrov (*Met. Fin.*, 1959, **57**, Oct., 72–73) An article translated from *Metallovedenie i Obrabotka Metallov*, 1958, Oct., 53–56, in which an investigation is described with the aim of establishing optimum conditions of electrolysis to ensure the formation of dense, fine-grained, and wear-resisting iron coatings, using chloride baths with various organic additives and under different conditions of temperature and current density.

Progress toward the development of a non-embrittling cadmium electro-plating process. II. Use of aliphatic amino-acids as complexing agents in aqueous baths P. N. Vlannes and S. W. Strauss (*Plating*, 1959, **46**, Sept., 1046–1051) The plating characteristics of aqueous Cd baths containing six aliphatic monoamino-acids have been investigated. The experiments are described in detail and the results discussed fully. Satisfactory plates were obtained from ammoniacal solutions of these baths, the results of delayed fracture tests indicate that these baths are less embrittling than the cyanide baths. **III. The use of triethanolamine in cadmium plating solutions** P. N. Vlannes, S. W. Strauss, and B. F. Brown (Oct., 1153–1157) A study of baths not containing cyanide is reported. With controlled current density embrittlement is not produced.

Finishing pointers: cadmium plating high strength alloy steels E. Calderon (*Met. Fin.*, 1959, **57**, Oct., 76) H_2 embrittlement is eliminated in Cd plating under the conditions described, using a KCN bath with low metal content and without brighteners, using vigorous agitation and high purity D.C. as free as possible from a.c. ripple. Parts are cleaned anodically, rinsed, and dipped in HCl containing no inhibitors. Other factors are described.

Effect of plating conditions on the structure and wear resistance of iron electrodeposits U. N. Petrov (*Electropl. Met. Fin.*, 1960, **13**, Jan., 21–23) A series of electrolytes were made up, all basically 500 g/l $FeCl_2.4H_2O$ and 80 g/l glycerol, in order to test the effect of various additives. The aim of the work was to obtain as sound, fine-grained and wear-resistant an electrodeposit as possible. The compositions and operating conditions giving the best results are given, and it is concluded that the results obtained are fully satisfactory and meet the requirements of the repair industry.

X-Ray investigation of the electrolytic deposition of manganese V. P. Moiseev and O. S. Ponova (*Zhurn. Fiz. Khim.*, 1959, (10), 2183–2189) Mn electrolytic deposits from sulphate baths with and without glycerine have been studied. Both the α- and the γ-modifications of Mn occur during electrolysis. The H_2 content of the precipitate and the kinetics of the phase transition during heating depend on the composition of the electrolyte. It is suggested that H_2 is present in the α-Mn as a solid solution.

Good looks and protection by metallising (*Engineering*, 1960, **189**, April 1, 438–439) A review of the process and products of vacuum evaporated coating.

When to consider automatic spray equipment W. Beach (*Met. Fin.*, 1959, **57**, Oct., 69–71, 76) Advantages of automatic paint spraying, suitable types of work, and equipment are considered, economic aspects being discussed.

Bonding evaporated metals to stainless steel A. S. Holik (*Met. Prog.*, 1960, **77**, April, 116) An electron micrograph (×15000) is reproduced to illustrate the fact that a stronger bond may be obtained on a rough surface. The technique used is described.

Antifriction phosphating of steel parts (*Metallrein. Vorbeh.*, 1960, **9**, April, 56–57) The phosphating of motor-car engine parts is described, including the pretreatment of the parts, from a Russian source. The solution is kept at 96–100°C so that very fine crystalline deposits are obtained which, after a chromate treatment and oiling, show exceptional antifriction properties.—T.G.

Investigating the phosphatizing of steel in the presence of nitrates of univalent metals I. I. Khain (*Zhurn. Prikl. Khim.*, 1959, **32**, (11), 2463–2468) Phosphatizing is a widely used anti-corrosion method. The nitrates of Na, K, and NH_4 added to the phosphatizing solution greatly influence some of the main factors of the process of formation of the phosphate film and its properties. In the presence of NH_4NO_3 the general acidity of the solution increases, the film formation is speeded up and its main properties change (thickness, coarseness, and shielding value against corrosion). The other nitrates have not the same effect on these factors. An explanation of this behaviour is suggested.

Investigating the phosphatizing of steel in the presence of nitrates of bivalent metals I. I. Khain (*Zhur. Prikl. Khim.*, 1959, **32**, (11), 2473–2479) The characteristic effect of the bivalent metal nitrates (Mn, Zn, Cd, Ca, Sr, and Ba) on the reaction is the speeding up of the process of film formation by Mn, Zn, and Cd, and its slowing down by the Ca, Sr, and Ba nitrates due to the increasing acidity of the solutions.

Phosphating, electropolishing and high speed chromium plating in buffer production H. L. Watts (*Electropl. Met. Fin.*, 1960, **13**, Jan., 15–19) A description of the application of these processes to a practical production problem.

A modern process for preparing sheet for painting: phosphating (*Met. Elect.*, 1959, **23**, July, 109–112) Description of phosphating using the Houghton product Houghton-Clean 313-A.

Chromatic and phosphate treatments for corrosion protection. 2 D. J. Fishlock (*Corros. Prev.*, 1959, **6**, Dec., 42–47, 58) The mechanisms of phosphating, measurement of coating thickness, and several types of coatings and applications are described.

Vitreous-enamelled cast iron J. R. Park (*Found. Trade J.*, 1960, **108**, April 7, 427–429) Illustrations of parts with notes on defects.

Developments in vitreous enamelling A. I. Andrews (*Vitr. Bull.*, 1959, **10**, Oct., 177–185) Steel composition and conditions promoting adhesion, theories of adhesion, relationships of firing conditions, enamel composition, processing, properties, tests, and control are discussed.

Interface phenomena in enamels F. Rickmann (*Vitr. Bull.*, 1959, **10**, Dec., 201–219) The mathematics of wetting phenomena and contact angles and their application in enamelling.

Diagnosis and cure of enamel faults D. L. Mill (*Vitr. Bull.*, 1959, **10**, Nov., 195–199) Faults occurring during the enamelling process are reviewed.

Pinholes and porosity of enamelled cast iron C. A. Sanders (*Vitr. Bull.*, 1959, **10**, Nov., 187–193) Factors are discussed and defects are illustrated.

Reaction kinetics of porcelain enamel—metal systems R. M. Spriggs and A. L. Friedberg (*J. Am. Ceram. Soc.*, 1960, **43**, May, 252–262) Powdered metals, including Fe, were reacted with borosilicate porcelain frits and the reactions studied by X-ray diffraction.

Shielding steels from gas corrosion by glass-metallic coatings E. A. Antonova and A. A. Appen (*Zhur. Prikl. Khim.*, 1959, **32**, (11), 2468–2473) The proposed glass–chromium coatings have considerable advantages over the silicate enamels because of their better properties and especially because of their superior shielding action and their property of resisting sudden changes in temp. They even surpass the metallic coatings formed by the diffusion process (thermo-chromiding and thermo-siliciding) in their heat resistant and shielding properties.

Testing of coal tar coatings. IV. Surface preparation and atmospheric exposure W. F. Fair jun. (*Corrosion*, 1960, **16**, Jan., 5t–8t) Large-scale test applications of various coal tar coating systems in a severely corrosive atmosphere demonstrate conclusively that good surface preparation such as by sand-blasting is of the utmost importance for long-time corrosion protection. The protective value of individual coal tar coatings depend also on the local corrosion conditions.—T.G.

Metal surface roughness: effect on enamel adherence A. Petzold and M. Kruger (*Ind. Fin.* 1960, **12**, March, 64–65) The authors have studied the effect of surface roughness on enamel adherence by subjecting samples of sheet to five different cleaning methods, applying enamel by a standardized method, and determining the adherence by three different procedures. The results clearly show that roughness and adherence increase together. The particular effects of pickling and shot-blasting are discussed.—G.F.

Metallic deposition by electrolysis: penetration Richter (*Galvano*, 1959, **28**, Oct., 459–464) Various methods of controlling and testing the thickness of penetration of electrolytic deposits are discussed and different factors affecting them are reviewed.—S.H.–S.

Organic additives: Choice of compound R. Glenat (*Ind. Fin.*, 1960, **12**, Jan., 35–39; from *Chim. et Ind.*, 1959, **82**, 329) The author indicates the effects of organic additives used in plating, and discusses how the molecular structure affects their influence on plating variables. The additives are grouped according to their utility, and the author discusses how each type can best be used.—G.F.

The plating of parts made of magnetic materials K. Sommer (*Metalloberfläche*, 1960, **14**, March, 84–86) Materials which have magnetic properties are often covered with very fine particles after machining or grinding. The effect of such particles on the plating of parts is discussed, and means are suggested for cleaning the surface.—T.G.

Porosity in electrodeposits: Fundamental considerations A. Kutzelnigg (*Metalloberfläche*, 1960, **14**, March, 81–84) After describing the various types of porosity in electrodeposits the author discusses the causes of porosity, the testing of porosity, e.g. by the ferroxyl test, and finally the effect of any pores on the corrosion-resistance of the deposit (28 refs).

Contribution to the study of the adhesion factor P. Badet (*Galvano*, 1959, **28**, Nov., 526–528) Low-carbon steels and cast iron are considered and various metal coatings are briefly reviewed.

Gas plating C. F. Powell (*Mat. Design Eng.*, 1960, **51**, Jan., 38–101) Five main advantages and three application methods and coating properties of gas plating are described. Its limitations at present are noted and future uses are suggested.—S.H.–S.

Depositing dense coatings of chromium, molybdenum and tungsten Alloyd Research Corp. (*Product Fin.*, 1960, **13**, Feb. 81), A

brief account of the vapour phase deposition method.

The development of chromium plating G. Dubpernell (*Plating*, 1960, **47**, Jan., 35–53) An historical review with portraits and coming up to the present time (196 refs).

Finishing satin chrome built-in ovens S. J. Beyer (*Plating*, 1959, **46**, May, 463–466) Plating and polishing techniques used for mass producing cooking-oven parts by G.E.C. are described. Bipolar and auxiliary anodes for plating, and automatic satin finishing are among the modern techniques used.—D.L.C.P.

Automatic plating of ironmongery Crompton & Sons Ltd (*Electropl. Met. Fin.*, 1960, **13**, March, 89) A description of a machine plating up to 30000 parts in 8 h with Ni and Cr.

Experiences with corrosion-resistant Inkrom steel screws E. Taubert (*Wire*, 1959, June, 60–62) The application of the Inkrom process for producing a Cr protective surface to screw manufacture, and the results of tests are described.

Finishing pointers J. B. Mohler (*Met. Fin* 1959, **57**, Dec., 74) Analytical and gravity control of Watts Ni baths are described.

The relationship between brightness and structure in electroplated nickel R. Weil and R. Paquin (*J. Electrochem. Soc.*, 1960, **107**, Feb., 87–91) Deposits from Watts baths were examined by electron microscopy.

Finishing small articles: Automatic plating. plant used (*Ind. Fin.*, 1960, **12**, March, 45–47) A new Efco-Udylite extended 'Junior' automatic plating machine has recently been installed at the Haycock plating factory of Thomas Crompton & Sons Ltd. The plant nickel and chrome plates up to 30000 components per shift, all cycles being accomplished automatically. The plant and its operation are described, and brief details are also given of the factory's other plating facilities.—G.F.

An outline of the chemistry involved in the process of catalytic nickel deposition from aqueous solution, Parts II, III G. Gutzeit (*Plating*, 1959, **46**, Nov., 1275–1278; Dec., 1377–1378); **Part IV** (1960, **47**, Jan., 63–70) The effects of the concentration of Ni, P_2O_2, and H-ions, of complexing and chelating agents and of stabilizers on plating rate and quality are discussed with notes on the Kanigen process.

Constructional form and hot-dip galvanizing W. Derenbach (*Metalloberfläche*, 1959, **13**, Nov., 357–359) Factors to be considered in the hot-dip galvanizing of parts of various sizes and shapes are reviewed.

Continuous galvanizing plant at August Thyssen-Hutte (*Engineer*, 1960, **209**, Jan. 15, 111) A brief illustrated account.

Galvanising sheet metal products J. H. Nichols (*Product Fin.*, 1960, **13**, Feb., 86–87, 124) Processes used by Garrods (Sheet Metal) Ltd are described.

Hot dip galvanizing practice (*Ind. Fin.*, 1960, **12**, Jan., 44–46, 48) A description is given of a new type of hot-dip galvanizing plant which combines a high-efficiency gravity-feed coke-fired furnace and an experimental bath setting. The results obtained on extended trials are compared with those of conventional practice. The new unit is considered suitable for continuous operation, but not for intermittent use.—G.F.

Zinc coatings on iron and steel, 4 finishes for zinc coatings A. K. Parker (*Met. Finishing J.*, 1960, **6**, Feb., 45–48) Chromate treatment and the use of organic finishes are surveyed.

Hot-dip galvanizing and paint—an ideal corrosion protection system K. A. van Oeteren-Panhäuser (*Metalloberfläche*, 1959, **13**, Nov., 360–363) The advantages of a combination of a hot-dip galvanizing layer with a coating of paint are discussed, and compared with a combination of paint and Zn coatings produced by other methods. Some operational data are given.

Electrolytic detinning (*Ind. Fin.*, 1960, **12**, March, 48) The New London Electron Works Ltd produces 25000-t of high-grade light steel briquettes per year from cleaned and detinned tin cans, at the same time recovering about 16 lb per t of Sn. Brief details are given of the electrolytic detinning lines and the method of operation.—G.F.

The electroplating of tin alloy solderable coatings on ferrous and non-ferrous basis metals A. M. Howard and L. R. Rogers (*Plating*, 1959, **46**, May, 484–487) Techniques of electroplating in preparation for soldering are reviewed. Various base metals and solder metals are considered. The value of 'reflowing' of the plated coating is explained.—D.L.C.P.

Electrolytic refining of black tin and heavy metal [$FeSn_2$] N. S. Golikov (*Stal'*, 1960, (2), 187–189) Electrolytic refining was not successful as porous, coarsely crystalline deposits were produced. Investigations gave a satisfactory bath composition with sulphamic acid and 2-naphthol and details of full-scale working are given. The method is much superior to heat refining.

Methods of testing the quality of tinplate W. E. Hoare and S. C. Britton (*Mitt. Forsch.-Blechverarbeitung*, 1959, Aug. 1, 209–218) Methods of testing thickness and uniformity of the tin coating, and weight of the oil film, corrosion resistance, and other tests are described.

Codeposition of tin–nickel plate from organic and mixed aqueous-organic solvents T. T. Campbell and R. Abel (*US Bur. Mines. Rept. Invest.*, 5482, 1959, pp.11) Less corrosive baths than the ammonium bifluoride type were sought. Triethanolamine-HCl and ethylene glycol-HCl baths were promising and had current efficiencies close to 100%. The deposit with 65% Sn–35%Ni could be obtained from organic or mixed organic aqueous-solvents.

Tin–nickel plating: Treatment and advantages (*Ind. Fin.*, 1960, **12**, Feb., 38–41; from *Tin & its Uses*, 1959, **48**, 1) The properties of tin–nickel electroplate, produced by the simultaneous deposition of equal atomic proportions of tin and nickel, are outlined and its corrosion resistance is compared with those of the two metals above. Details are given of the plating procedure.—G.F.

Progress in the development of a non-embrittling cadmium electroplating process P. N. Vlannes, S. W. Strauss, and B. F. Brown (*Plating*, 1959, **46**, May, 467–468) Progress is described in developing a Cd plating process which will avoid the H_2 embrittlement of steel encountered with the cyanide bath. It is concluded that solutions of glycinate, methanol, and triethanolamine are less embrittling than cyanide and warrant further investigation.

New coating systems, principles and application W. Singhof (*Mitteilungen der Forschungsgesellschaft Blechverarbeitung e.V.*, 1959, July 1, 185–190) Modern lacquers on chlorinated rubber, epoxy, polyester, and other bases are described and their applications indicated.

Surface protection by fluidized powder sintering and flame spraying with plastic powders E. Gemmer (*Schweissen Schneiden*, 1959, **11**, Oct., 389–393) The techniques of plastic coating by the two processes are described, several examples are given, and the available plastics and their resistance to chemical attack are reviewed.

Plastics for corrosion protection in the chemical industry A. R. Gabel and H. W. Schmidt (*Chem. Eng. Progress*, 1959, **55**, Nov., 39–41) A review.

Fluidized bed applies plastic coating W. M. Stocker (*Metalw. Prodn.*, 1960, **104**, April 6, 622–623) A method for the plastic coating of wire products with vinyl powder or similar materials is indicated.

Fluidized bed coating—solution to difficult finishing operation W. R. Pascoe (*Met. Fin.*, 1959, **57**, Oct., 74–76, 80) A method is described of vinyl-coating a welded wire rack, by immersing the heated part in a fluidized bed in the finely-divided powder. No deburring, grinding, or polishing of the rack is required.

Promising precoated strip is a teaser (*Engineering*, 1960, **189**, March 18, 374–375) A review of advantages and a table of products, manufacturers, and characteristics. Joint designs are surveyed and an outline of production is added.

Notes on the painting of steel structures N. A. Bennett (*Corros. Prev.*, 1959, **6**, Dec., 51–52, 56) Importance of accessibility of structures for painting, pre-treatment, and minimum acceptable thickness of coating are discussed.

The surface preparation of ship plate (outer bottom plating) for painting J. C. Hudson, J. F. Stanners, and A. G. B. Miller (*Trans. Inst. Marine Eng.*, 1959, **71**, Sept., 269–298) Raft tests are described confirming that compositions may not adhere to freshly descaled unrusted steel.

CLAD SHEET AND HARD-FACING

Hard surfacing of machine parts and tools with the welding torch H. Killing (*Schweissen Schneiden*, 1959, **11**, Nov., 432–436) Techniques and applications are described, and several examples of finishing and repairing are given.

Automatic hard facing on a rotating bench of the large cone of the blast furnace charging installation M. A. Tylkin, B. I. Sivak, J. F. Parfent'ev, and M. A. Kropp (*Avtom. Svarka*, 1959, (9), 88–93) A stand is described on which is mounted the large cone of the blast furnace top and which rotates under a welding machine.

Hardsurfacing improves wear resistance C. E. Raymond (*Canad. Metalw.*, 1960, **23**, April, 32–34) Hardfacing alloys are classified and various properties and reasons of choice are discussed.

Practical aspects of hard chromium plating C. F. Corfe (*Electropl. Met. Fin.*, 1960, **13**, Jan., 3–8) The properties of hard-Cr plate, the plant required to deposit it, and the conditions for satisfactory deposition are discussed.

Steel bolts protected against corrosion G. Becker (*Maschinenwelt Elektr.*, 1959, **14**, 19–20, 482–483) 'Inkrom' steel bolts and setscrews are given a Cr coating, diffused into the surface, by means of vapourized halogen compounds.

Practical lower-cost process diffuses chromium into steel Alloy Surfaces Co. (*Iron Age*, 1960, **185**, Feb. 11, 143–145) The 'Alphatizing' process for giving steel parts a chromium alloy surface is described. The process involves heating the parts with the material 'Alphalloy' in a suitable furnace. Besides giving a stainless surface, steels so treated have good formability and weldability and high wear resistance.—D.L.C.P.

Stainless steel clad dispersion of boron in iron for pressurized water reactors C. F. Leitten jun., R. J. Beaver, and A. E. Richt (*J. Nuclear Mat.* 1959, **2**, July, 136–144).

POWDER METALLURGY

Forms and shapes of materials: metal powder parts (*Mat. Design Eng.: Materials Selector Issue* 1959–60, 1959, **50**, 360–361) Properties and design factors are briefly surveyed.

A new method for compacting metal or ceramic powders into continuous sections F. Emley and C. Deibel (*Metal Powder Ind. Fed. 15th Annual Meeting*, 1959, 5–13) A shaped punch method is described.

Strip made from ore at new research centre (*Blast Furn. Steel Plant*, 1959, **47**, Aug., 857–858) In the new Independence, Ohio, research centre of Republic Steel Corporation, ore is processed to strip on a laboratory scale without melting. Low grade iron ores are reduced with coke fines and a purified iron powder produced which is rolled to strip.

Slip casting L. M. Schifferli jun. (*Met. Prog.*, 1959, **76**, Oct., 99–102) An outline is given of the technique of slip casting, and the adaptability of the process is illustrated by several examples.

Slip casting technique widens powder metallurgy H. H. Hausner and A. R. Foster (*Steel*, 1960, **146**, Feb. 15, 120–123) A slip-casting technique similar to that used for ceramics is described for metal powders. Metal powder with carefully chosen particle sizes is made into a slip and cast in a mould made of plaster of paris or other porous material. After solidification of the slip the mould is removed and the part dried and sintered. Hollow shapes can be obtained by drain casting. Very high densities, as high as from pressed powders can be obtained. The saving on presses and metal dies may make the method economical for short runs. Stainless steel parts have been made from powders of the 30 to 100 μ size range.

Detergency during infiltration in powder metallurgy T. Kimura, J. C. Kosco, and A. J.

Shaler (*Metal Powder Ind. Fed., 15th Annual Meeting*, 1959, 56–66) Infiltration and detergency theory is outlined and tests with Fe powder using various materials as solid, liquid, or vapour are described.

The physics of the sintering process on carbonyl iron H. D. Dietze (*Techn. Mitt. Krupp*, 1959, **17**, (3), 103–110) Sintering tests were made on pressed carbonyl iron in the α- and γ-zones; these showed a linear relationship between the porosity of the specimens after pressing and after sintering. Activation energy for the density increase in the α- and γ-zones was found to be 40 kcal/mol and 45 kcal/mol respectively. The experimental method of determining the mean pore surface curvature from metallographic sections is described.

Some sintering atmospheres obtained with incomplete combustion of methane used in ferrous powder metallurgy (*Rev. Roumaine Met.*, 1959, **4**, (2), 169–179) [In Russian] A wide range of gas mixtures may be obtained by incomplete combustion of CH_4. The effect of these gases on the compacting of iron powder subjected to sintering shows that over a wide range of concentrations corresponding to a cracking giving a ratio of excess air between 0·45 and 0·65, sintering occurs without carburizing or oxidation. The gas mixtures are reducing and the inspection of the surfaces as well as of the microstructures does not reveal any traces of oxides.

Contribution to the study of the sintering phase of sintered hard metals M. Petrdlik and V. Dufek (*Hutn. Listy*, 1959, **14**, (9), 786–790) Examples are given of the attainment of the optimum state of bonding of different component particles of sintered hard metals to obtain the most satisfactory mechanical properties (36 refs).

Accurate regulation of sintering temperatures simplified by power input control system Fansteel Metallurgical Corp. (*Indust. Heat*, 1959, **26**, Oct., 2000–2002) A system for rapid controlled heating has been installed in Fansteel Metallurgical Corporation's plant, which sinters electrical contacts in H_2 at temp. in the range 1000–1400°F to 1800–2500°F. Each furnace has two zones, controlled by Speedomax H controller-recorders, with integral current adjusting control for driving saturable core reactors; the input is varied according to the size, duration, and speed of temp. deviations.—K.E.J.

Permanent magnets made from fine iron powder A. S. Eisurovich (*Izvest. AN Metallurg i Topl.*, 1959, (4), 130–134) Formates of Fe, Co, Ni, and other metals are now supplied by the chemical industry. The main operations in the manufacture of these magnets consist in the reduction of the iron formate (or salt mixtures) and compacting the magnets from the powders with subsequent annealing of the pieces. To avoid ignition, the powders are moistened with benzene and then dried.

The development of 50:50 nickel iron for magnetic-amplifier cores G. P. J. Buchi, T. B. Burnett, and J. E. Thompson (*Powder Met.*, 1959, (4), 15–22) The production of 50:50 Ni–Fe strip by powder metallurgy techniques is described. A method of testing the suitability of this strip was developed, and is described. It was found that annealing twins adversely affect the properties of the strip, and the correct rolling and annealing conditions to avoid twin formation were determined.

The production of grain-oriented 50:50 nickel iron magnetic strip by cold rolling from sintered compacts E. V. Walker and R. E. S. Walters (*Powder Met.*, 1959, (4), 23–31) The sintering, soaking, final annealing, and final cold reduction conditions required to achieve the required preferred orientation are discussed, and the magnetic properties resulting from the recommended treatment are given.

The production of silicon iron magnetic strip with the (110) [001] texture, by cold rolling from sintered compacts E. V. Walker, J. Howard, and D. S. Shotton (*Powder Met.*, 1959, (4), 32–34) The sintering conditions required to produce the (110) [001] texture in Si–Fe strip are described, and the magnetic properties of oriented and non-oriented strip are compared. In an appendix an X-ray method is described.

The fabrication of soft magnetic alloy strip from powders H. H. Scholefield and S. G. G. Richardson (*Powder Met.*, 1959, (4), 44–56) The alloys investigated were mumetal, Ni–Fe alloys, Si–Fe, and Co–Fe. The techniques employed to produce the various types of strip are described. The results of varying the compositions of these alloys are given, and the role that powder metallurgical methods have to play in the production of soft magnetic alloy strip is discussed.

The variation in intrinsic coercivity of iron–cobalt powders prepared by electrolysis into mercury W. Wright (*Powder Met.*, 1959, (4), 79–89) The intrinsic coercivity of Fe, Co, and Fe–Co powders prepared by electrolysis was measured at −60°C, and found to be affected by heat-treatment, and by stirring of the (Hg) cathode during electrolysis. The effect of Sn additions was also investigated.

Use of powder metallurgy for diffusion bonding of cemented carbide to steel J. F. Kuzmick and G. B. Goodfellow (*Metal Powder Ind. Fed., 15th Annual Meeting*, 1959, 14–21) An account of 'Plymet' self-fluxing wafers and their uses.

The coercive force of fine iron powders E. H. Carman (*Powder Met.*, 1959, (4), 1–14) The dependence of the intrinsic coercive force of compacts prepared from H_2-reduced Fe_2O_3 powder on particle size has been investigated for particles in the range 200–800 Å. It was found that the coercive force reached a peak value of 970 Oe when the nominal particle size was 260 Å. It was shown by X-ray line broadening studies that packing effects, rather than strain anisotropy within the particles, control the coercive force of compacts. The agreement of the results with the theories of Néel, Stoner, and Wohlfarth and Jacobs and Bean is discussed.

On the reactivity towards water of iron powders obtained by reduction of ferric oxides P. Dugleux and J. Hui (*Compt. Rend.*, 1959, **249**, Nov. 9, 1898–1900) In the absence of O_2 the reaction with water is a sensitive test for reactivity in powders reduced at low hydrogen pressure and low temp.

Direct measurement of fine metal powders. Evaluation of micromesh sieves from 10 to 45 microns J. Haertlein (*Metal Powder Ind. Fed. 15th Annual Meeting*, 1959, 49–55).

CARBIDES, FERRITES AND CERMETS

Properties of ferrites and their application to electronics M. A. Michel (*Ind. Nat.*, 1960, **1**, 11–25) After giving a short historical survey, the three main types of ferrites are discussed: spinels of general formula Fe_2O_3, MO where M is a bivalent metal; ferrites with hexagonal structure, 6 Fe_2O_3, MO where M^{2+} is greater than 1 Å; rare earth ferrites 5 Fe_2O_3.3M_2O_3 where M is a tervalent rare-earth metal. The physical and thermal properties of ferrites are dealt with in detail and their applications are discussed in the light of these properties.—R.P.

Formation of ferrites during the ageing of metal hydroxides V. P. Chalyi (*Ukrainsk. Khim. Zhur.*, 1959, **25**, (3), 285–287) X-Ray structure analysis has shown that during ageing under water at room temp. of binary metal hydroxide systems, interaction takes place between the hydroxides with the formation of chemical compounds of the spinel type of structure (ferrites). In these binary systems the formation of Ni, Mg, and Zn ferrites was observed, but not that of Cu and Cd.

Ferromagnetic crystalline anisotropy of Fe Mn-ferrite system Z. Funatogawa, N. Miyata, and S. Usami (*J. Phys. Soc. Japan*, 1959, **14**, Nov., 1583–1587) In this work the dependence of the magnetic crystalline anisotropy on composition was investigated. The compositions used were in the range $Mn_x\ Fe_{(3-x)}O_4$ $(0{\cdot}2 \leqslant x \leqslant 1{\cdot}15)$.

Dry pressing of barium ferrite in the orienting field M. Nagakura and T. Yamaguchi (*Nippon Kinzoku*, 1959, **23**, July, 373–376). **Magnetic domain patterns on barium ferrites** (407–409).

A study of the magnetic properties of sintered ferrites, using single crystal data F. W. Harrison, K. Hoselitz, and J. E. Knowles (*Powder Met.*, 1959, (4), 90–104) The effect of the conditions of preparation of ferrites on their magnetic properties is indicated. The necessity for data obtained from single crystals, in order to design and understand the properties of polycrystalline ferrites, is explained, and methods of preparation of single crystals are described. The use of such data is illustrated by reference to a 'square-loop' ferrite.

The ageing of permeability in manganese zinc ferrite S. Miyahara and T. Yamadaya (*J. Phys. Soc. Japan*, 1959, **14**, Nov., 1635) The effect of thermal and A.C. demagnetization of ferrite specimens on their permeability was investigated. From the results it is deduced that decrease of permeability after sintering is due to disaccommodation, and not to irreversible structure changes.

The effect of grain size on saturation magnetization of barium ferrite powders K. Torkar and O. Fredriksen (*Powder Met.*, 1959, (4), 105–107) The authors have shown that the saturation magnetization of these powders decreases with decreasing grain size. This decrease is attributed to the existence of a disturbed layer on the surface of the individual grains (Huttig's surface layer).

A study of the systems Fe_2O_3 La_2O_3 and Fe_2O_3 Sc_2O_3 J. Cassedanne and H. Forestier (*Compt. Rend.*, 1960, **250**, April 25, 2898–2900).

Magnetic ceramics. VII. Effect of iron oxide particle size on nickel ferrite formation G. Economos and T. R. Clevenger jun. (*J. Amer. Ceram. Soc.*, 1960, **43**, Jan., 48–52) The extent of nickel ferrite formation as influenced by the particle-size distribution of the iron oxide constituent was measured and the activation energies calculated for increasing particle size. The results are examined and compared with current procedures for the production of ferrite powders.—T.G.

Iron alumina materials A. Gatti (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 753–755) Al_2O_3 was dispersed in Fe by various methods, powdered mixtures were compacted into rod, and microstructures and some mechanical properties examined, and the data are presented.

Metallo-ceramics of iron and steel S. A. Tsukerman (*Vestnik AN SSSR*, 1959, **29**, (9), 42–45) A plea for the wider use of powder metallurgy in the USSR.

Magnetic properties of sinters of different degrees of fluxing, as depending on the content of ferrous oxides N. A. Gladkov, Z. I. Nekrasov, and V. V. Chekin (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 42–44) Magnetization and the coercive force of sinters of different degrees of fluxing were measured and a region has been established in which variations in the basicity made almost no impression on the magnetic properties of the ferromagnetic components. The conditions for the formation of the sinters in this zone were almost similar, but differed from those of unfluxed sinters and from sinters with a high degree of fluxing corresponding to a basicity of 1·8.

Manufacture and properties of metal ceramic iron nickel aluminium magnets P. A. Gladyshev and V. L. Memelov (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 106–110) Magnets made from Alni, Alnico, and Magnito alloys give complete satisfaction in the home manufacture of electrical apparatus, relays, spark extinguishing, and magneto-electric devices etc. The methods of production are adaptable for the manufacture of small magnets of intricate design in weights from fractions of a gram to some tens of grams.

PROPERTIES AND TESTS

Metallurgical research in Great Britain N. P. Allen (*Metallurgist*, 1960, **1**, May, 106–107) A brief account of the types of bodies carrying out research.

The examination and testing of metallic materials by physical methods H. Thompson (*Inst. Auto Assessors J.*, 1959, **10**, April, 29–47) Impact, creep, fatigue, crack detection, and ultrasonic, magnetic, and X-ray tests are outlined with illustrations of apparatus (40 refs).

Study on high-purity pig iron. I. II. Studies on the special features of high-purity pig iron. Parts 1 and 2 K. Takai and Y. Sato (*Tetsu to Hagane*, 1959, **45**, Sept., 899–901; 901–903) I. Detailed analyses, for up to 15 elements, are given for seven irons (including charcoal irons)

having total impurities in the range 0·089–0·793%, exclusive of C, Si, and Mn. Some stress-strain diagrams are given. II. The mechanical properties of the spheroidal graphite irons after heat-treatment are given, and the effects of P on mechanical properties and length of cracks are shown.—K.E.J.

Mobile low-temperature apparatus S. S. Chang and J. D. Isdale (*Metallurgia*, 1960, **61**, Jan., 6–8) Test equipment for holding test specimens at constant sub-zero temp. for up to 90 min is described. Temp. down to −39°C have been so maintained and −80°C has been reached.

On the properties of several kinds of rails Y. Koike, Y. Simizu, and M. Sasaki (*Tetsu to Hagane*, 1959, **45**, Sept., 982–984) Chemical composition, tensile properties, abrasion-test results, and S–N diagrams are given for C, Mn, and Cr steel rails.—K.E.J.

Surface finish measurement reviewed B. Piper and W. P. Callis (*Metalw. Prodn.*, 1960, **104**, Jan. 6, 13–25) Commercial instruments and surface finish standards are extensively reviewed.

On model studies of metallic surface asperities F. F. Ling and R. C. Lucek (*J. Appl. Phys.*, 1959, **30**, Oct., 1559–1563) Sliding contact between the tips of cones is measured with or without lubricant, and with change of cone angle and discussion of the size effect.

Regarding the specified limits of roughness R. Motoiu (*Met. Constr. Masini*, 1959, (11), 950–958) The tolerance in the execution of a mechanical operation is determined by the relation T = a.i. where a is the number of tolerance units of the respective class of accuracy and i the unit of tolerance which according to Rumanian standards is $i = 0{\cdot}5\ \sqrt[3]{D}\ (\mu)$ where D is the arithmetical mean of the initial and the final diameters of the workpiece. The relation between roughness and tolerance is given by R = f(T) = f(a.i.) and the general form of this expression is $R = k.a^n.i^m$ where K is a constant and m is the index for the effect of the diameter on the roughness and n the index for the effect of the class of accuracy on roughness. If therefore K < 1, the roughness will be smaller than the tolerance.

An introduction to theories of plasticity R. Epain and J. Kieffer (*Mém l'Artillerie Franç.*, 1959, **33**, (4), 927–964) A detailed review.

The effect of an ultrasonic field on the elastic properties of steel during zinc plating A. Ginberg and A. P. Gorina (*Metalloved. Term. Obra. Met.*, 1959, (8), 20–24) During chemical degreasing in a hot solution containing 100 g/l NaOH a certain reduction of elastic properties of U8A steel is observed. On applying an ultrasonic field, the reduction of the elastic properties increases at first, to regain thereafter its initial value. Ultrasonics speed up etching in HCl containing 8–10 g/litre inhibitor PB–8 by a factor of 3–5.

The effect of cold plastic deformation on the modulus of elasticity of carbon steels A. V. Gur'ev (*Fiz. Met. Metalloved.*, 1959, **8**, (5), 777–784) After plastic deformation the changes in the true value of the modulus of elasticity E = f(ε) and of the shear modulus G = f(γ) where ε and γ are plastic deformations) can be divided into two zones; one of low stresses showing a sharp decrease in E and G and the other, a zone of medium and high stresses, showing a linear decrease in these moduli. This points to the fact that in the second case the intensity of the transition to plastic deformation of the microvolumes becomes stabilized and then remains constant while the sharp increase in the first zone may be explained by the weakening of the layer in the specimen in which plastic deformation may originate under low stresses.

The effect of temperature on the relaxation and plasticity of quenched steels S. S. Shurakov (*Metalloved. Obra Met.*, 1959, (7), 10–15) The relaxation processes at the grain boundaries of a non-stressed specimen all things considered, may take place because of individual displacements of atoms or because of the diffusion of the iron atoms. According to the data in the literature the activation energy of self-diffusion varies between 47·4 and 77·2 kcal/g atm, which means that it exceeds appreciably the value of the activation energy of the relaxation of quenched steel which is 9 kcal. This low activation energy of relaxation may be due to the fact that self-diffusion of iron occurs only in localized zones at the boundaries and also because of the particularly sharp distortions due to the special features of the martensite reaction.

The fine structure of brittle and fibrous fractures of the EI 531 steel tubing in expansion tests with a cone E. Yu. Chemadurova (*Fiz. Met. Metalloved.*, 1959, **8**, (3), 466–468) The tests were carried out in the cold with a mandrel with a 1:10 taper after heat treatment, up to an increase of 6% of the external dia., when no cracks, tears, laminations, or laps should occur, and also by expanding up to failure. From a comparison of the data obtained regarding microstructure, hardness, distortion of the second order, and blocks, it is possible to conclude that the different types of fractures are related to the dimensions of the blocks.

Aspects of theories of dislocation mobility and internal friction J. Lothe (*Phys. Rev.*, 1960, **117**, Feb., 704–708) Theories of kilocycle-range internal friction are discussed.

Low temperature internal friction in FCC and BCC metals L. J. Bruner (*Inst. Study Met. Univ. Chicago*, 1959, Dec., 545–573) Fe is included.

Effect of vanadium, tungsten, chromium, and molybdenum on internal friction and rate of ageing of technical iron M. G. Lozinskii and A. E. Fedorovsky (*Izvest. AN Metallurg. i Topl.*, 1959, (1), 64–70) It may be concluded on the basis of the extensive experiments described that the effect of V consists of binding the N atoms, preventing their free displacement in the iron lattice with changing stresses. No final conclusions can be reached regarding the effect of V and Cr on ageing. W additions increase slightly the rate of decay of the solid solution in the second stage of ageing and considerably increase the rate in the first stage.

Observations of dislocations in metals by transmission electron microscopy P. B. Hirsch (*J. Inst. Met.*, 1959, **87**, Aug., 406–418).

The measurement of grain contiguity in opaque samples J. W. Cahn and J. E. Hilliard (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 759–761) It is suggested that degree of contiguity, or contact, between adjacent grains be described by three parameters – the total area, S, per unit volume of contacts between grains of specified phases, the total edge, λ, in a unit volume between three grains of specified phases, and the total perimeter C of contact areas in a unit volume between specified phases. These parameters can be determined rigorously from measurements on a random plane of polish. An expression is derived for number of contacts per unit volume.

On the contrast of the electron microscope image due to an edge dislocation H. Hashimoto nd M. Mannami (*Acta Cryst.*, 1960, **13**, April 10, 363–364) Calculations are summarized and formulae given.

Testing for notch sensitivity in welded joints C. E. Hartbower (*PB* 151630; pp.24) Conditions to be observed for obtaining a satisfactory result in using the V-notch Charpy test are discussed, the test being applied to weld metal, heat-affected zone, and base metal.

The photoelastic picture of shock waves in the elastic half-plane as a function of shock duration and shock velocity H. Lange and W. Habbel (*Forsch. Nordrhein–Westfalen*, 1959, (719), pp.56) The propagation of shock waves produced by falling weights, bullets, and explosive charges in a photoelastic material is derived from the photographs of the isochromates and isoclinics. The method is explained and the results are interpreted by the use of models.—T.G.

The relationship between the deformability of steel and its structure N. S. Alferova (*Stal'*, 1960, (2), 144–148) The form of the curves relating plasticity to temp. is discussed. The concept of 'equal tensile strength temp.' would require the curve to show a single maximum but in fact there are various types of curves. These are now studied by referring to transformation temp. and metallographic structures. Hot and cold rolling conditions are considered.

Strain-hardening solutions to axisymmetric discs and tubes N. Perrone (*Trans. ASME J. Appl. Mech.*, 1960, **27E**, March, 45–53).

The special features of the gradual uniform plastic deformation of steels. II L. G. Afendik (*Fiz. Met. Metalloved.*, 1959, (6), 910–914) The quasi-isotropism in the mechanical properties of steels in plastic deformation is observed only with uniform deformation. A disturbance in the uniformity of the change in the macrodisplacement results in the appearance of deformation anisotropy. One of the principal features of the gradual uniform plastic deformation is the reduction in the resistance to such deformation. Relaxation and ageing reduce the tensor components of anisotropy.

Studying the failure of steel from the formation of the fracture E. M. Shevandin, V. M. Dudasheva, and R. E. Reshetnikova (*Fiz. Met. Metalloved.*, 1959, (6), 922–928) Experimental curves have been constructed for the relation between the dimensions of the average, the largest, and the smallest facets of coarse-grained carbon steels and temp. The melting processes, as disclosed by the curves, the presence of crystalline facets appearing with the slight magnification of the fibrous fractures and the fact that these facets stiffen before the appearance of the zone of crystalline fracture; confirm the opinion of the dominant role played by the normal stresses in the failure of metals, both brittle and plastic.

Dynamic and static deformation of Armco iron H. P. Tardif and P. Chollet (*Nature*, 1959, **184**, Dec. 19, 1917–1920) Specimens were deformed statically and by impact and then cut in two and one-half aged. Both were then tested for hardness. The increase in hardness was less for the dynamically deformed specimens. After ageing the difference in hardness was again approximately constant. This held except at −196°C, but even then the hardness after dynamic deformation was lower. Dislocation theory is invoked in explanations.

Study of the strain-ageing of mild steel and of high-purity iron by age-hardening and reversion tests M. Nacken and W. Heller (*Arch. Eisenh.*, 1960, **31**, Feb., 103–111) The course of strain-ageing was followed in high-purity iron, converter, and OH steel, and electro-steel by measurement of the damping capacity as an indication of the reduction of the dissolved-carbon content during age-hardening at 20–140°C. The course of ageing obeyed in all cases the Cottrell–Harper precipitation law. These results were supplemented by reversion tests. The experiments are described in detail and the results presented in many graphs and tables.—T.G.

Effect of stress concentration on the fatigue resistance of steel E. M. Shevandin, L. F. Kurdyukova, and L. M. Rubinshtein (*Fiz. Met. Metalloved.*, 1959, (1), 122–129) With an increase in the acuteness of the notch and in the stress concentration, fatigue strength of low carbon and low alloy steels decreased both in bending and in tension-compression, and falls to a minimum value at which even with further increases in the acuteness of the notch, it remains stationary. The radius of the notch corresponding to this limiting value of fatigue strength may be called the limiting radius of the notch. So for example, for specimens under bending and tension-compression tests, 30 to 60 mm² in cross-section, this radius has a value between 0·2–0·5 mm. A gradient in the stress conditions increases fatigue strength in bending as distinct from tension-compression, but this effect shows preferentially in smooth specimens.

Theoretical and actual stress-concentration factors in fatigue tests of metallic materials J. M. Pintado Fe (*Tecn. Met.*, 1958, **14**, Sept.–Oct., 165–174) [In Spanish] The influence of stress raisers upon fatigue strength is considered and work done upon this topic is reviewed. Theoretical and actual stress concentration factors and the notch sensitivity of the material are defined and curves showing how these vary with the condition of the material are given for a Spanish national steel, F–123, and discussed. The various theories that have been proposed for deducing the notch effect are examined but it is thought that experimentally determined curves of the type discussed would provide more useful evidence to the designer.

The fatigue strength of some steels at low temperatures H. Hildesheimer (*Schweiz. Arch.*, 1959, **25**, June, 187–200) Four ferritic steels were investigated (plain C or low alloy) and an 18/8 steel, at temp. of 4·2, 84 and 291°K; the test machine used is described, and the results discussed (24 refs).

Increase in fatigue resistance by local heating of plates and constructions with welded seams O. Puchner (*Rev. Soud.*, 1960, **16**, (1), 28–34) The fatigue limit of welds when measured proved to be lower than that of the base metal. Heat-treatment produces changes in the stress systems especially under compression and processes have been developed raising considerably the fatigue strength of the welds.

Cumulative fatigue damage at elevated temperature W. K. Rey (*NACA TN* 4284, 1958, Sept., pp.53: from *US Res. Rep.*, 1959, **31**, March 13, 163) Heat treated SAE 4130 was examined at room temp., 400°, and 800°F. The results are compared with theory.

The fatigue strength of chromium plated steel T. C. de K. Knun (*South African Mech. Eng.*, 1960, **9**, Jan., 139–144) Rotating bend fatigue specimens with and without 0·005 ± 0·002 in. plating were tested. Complete and partial replating after simulated wear was also carried out. The effect was smaller but in the same direction as trends found in harder steels.

Fatigue studies of plated hot work tool steels G. W. Lawless, V. F. Lardonoit, and B. Cohen (*Plating*, 1959, **46**, Nov., 1264–1267) Cr-plated steels are developed for aircraft use up to 1000°F, these require protective coatings. Fatigue strength is greatly reduced by plating but work is described to obviate this. Shot-peening and nitriding before plating largely restore the original endurance.

Crack propagation tests of some high-strength sheet steels J. E. Srawley and C. D. Beachem (*NRL Rep.* 5263; *PB* 151352, 1959, Jan., pp.26; from *US Res. Rep.*, 1959, **13**, April 17, 249) Eight steels were examined over a range of temp. A tentative criterion for evaluation is suggested related to service temp. Many factors are discussed and two of the steels tested combined a room-temp. yield strength >200 psi with a fracture-appearance transition temp. at or below room temp.

Fatigue strength of cylinders with cross bores J. L. M. Morrison, B. Crossland, and J. S. C. Parry (*J. Mech. Eng. Sci.*, 1959, **1**, Dec., 207–210) A survey of theoretical analyses and practical tests with summarized results.

Propagation of fatigue cracks in various sheet materials N. E. Frost (*J. Mech. Eng. Sci.*, 1959, **1**, Sept., 151–170) Steels and non-ferrous metals were tested using plates with a slit to initiate fatigue cracks. The results were analysed on the hypothesis of Frost and Dugdale and a growth law was formulated.

The failure of structural metals subjected to strain-cycling conditions R. W. Swindeman and D. A. Douglas (*Trans. ASME (D)*, 1959, **81**, June, 203–212) An investigation of the effect of strain-cycles on Inconel, Hastelloy B, and Be is reported. The relationship $N^{\alpha}\varepsilon p = K$ is verified for the metals at elevated temperatures providing the time cycle is relatively short (N = no. of cycles to failure; εp = plastic strain per cycle; α and K = constants depending on the material and test conditions). It was found that above 1300°F grain-size and frequency exert a pronounced effect on the rupture life. Fine-grained metal survives more cycles before failure than coarse grained material. Long time cycles shorten the number of cycles to failure when the strain per cycle is low. Good correlation was found between the effects of thermally and mechanically produced strains.—D.L.C.P.

Impact characteristics of heat-affected zones in Mn Mo armour steels E. F. Nippes, W. F. Savage, and J. M. Paez (*Weld. J.*, 1960, **39**, Jan., 31s–39s) Four steels were studied in order to determine the effects of welding thermal cycles upon the notch toughness of the partially transformed regions of the heat-affected zone. It was found that heating to between Ac_1 and Ac_3 (1350–1600°F) produced banded, multiplex structures, and that the transition temp. (based on the Charpy V-notch 10 ft/lb criterion) was markedly raised in every case. On exposing this zone to a further thermal cycle with a peak of 1000°F, the impact properties were entirely restored.

Effect of microstructure on notch toughness —Part IV E. H. Kottcamp jun. and R. D. Stout (*Weld. J.*, 1959, **38**, Nov., 435s–440s) Al-killed low-carbon plain and alloyed steels were used in this investigation. The object was to test whether ferrite grain size alone was the controlling factor in affecting notch hardness. It was found to be of primary importance, but not to be the only factor. The effects of the various alloying elements on notch-toughness after air-cooling, furnace cooling, or one of these followed by tempering for varying times was investigated. The site of fractures in ferrite–pearlite aggregates was also investigated, and confirmed the views of Danko and Stout.

Conchoidal fractures in nickel vanadium steels Z. Eminger (*Neue Hütte*, 1960, **4**, Oct., 596–608) Influence of composition, particularly S, Al, and Mo in the as-cast condition, of heat-treatment and working, state of deoxidation and gas content on the incidence of conchoidal fracture is examined with reference to the literature and to a large number of experiments (23 refs).

Features of the plastic deformation of technical iron B. S. Kasatkin (*Izvest. AN Metallurg i Topl.*, 1959, (5), 59–69) On the basis of electron microscope studies of this deformation one may conclude that in the early stages of plastic deformation slipping within the grains is insignificant but it does occur in a large number of planes. This results in the formation of the A system of slipping. Optical microscopes have not sufficient resolution for the accurate observation of these weak slip lines. In the later stages of plastic deformation the intensity of the original slip lines increased and their number multiplies rapidly.

Study on high-manganese steel. II. Effect of carbon and manganese contents on the plastic deformation of high-manganese steel. III. Effect of silicon, copper, and molybdenum contents on the plastic deformation of high-manganese steel Y. Kido and H. Fujii (*Tetsu to Hagane*, 1959, **45**, Sept., 1097–1098; 1099–1100).—K.E.J.

The effect of the rate of heating and cooling on the irreversible deformation of iron S. F. Kovtun (*Fiz. Met. Metalloved.*, 1959, **8**, (6), 939–945) The thermal changes of cylindrical specimens of iron, St3, St40, U8A, and U12A steels were studied *in vacuo* by means of repeated successive heating and cooling cycles in a range of temp. which included the $\alpha \rightleftarrows \gamma$ points. An appreciable residual change in the length of all specimens could be observed. This irreversible change is due to phase transformations and may be reduced by admixtures, especially carbon. The rate of heating and cooling is the principal factor in these changes under repeated thermal cycles.

The effect of interposed lubricants on the surface deformation of metals during plastic working L. H. Butler (*J. Inst. Met.*, 1960, **88**, April, 337–343) Mild steel was included in the tests. A mathematical treatment is given in the appendix.

The problem of the anisotropy of metals after plastic deformation. The method of determination of the degree of anisotropy in polycrystalline metals W. Truszkowski (*Acta Met. Sin.*, 1959, June, **4**, 127–134) [In English] A proposed method of determining the coeff. of anisotropy of cold-worked metals is based on measuring the deformation in the neck of a strained test-piece at various cross-sections and extrapolation to the region of uniform elongation. This is confirmed by results on Ni which is annealed and deformed by cold rolling to different degrees.—K.E.J.

The work hardening of metals in industrial processes of plastic deformation W. Truszkowski (*Acta Met. Sin.*, 1959, June **4**, 135–142) [In English] Work-hardening curves can be obtained on the basis of a single tensile test on a metal specimen by using the correct relationship to calculate σ_i/σ_{pb}, the ratio of true stress to yield stress in the region of necking of the specimen (14 refs).—K.E.J.

Failure by fatigue L. F. Coffin jun. (*GEC Research Lab. Bull.*, 1956–60, Winter, 26–27) Testing of tubular specimens is briefly described.

The cyclic straining and fatigue of metals L. F. Coffin jun. and J. F. Tavernelli (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 794–807) Deformation and fracture characteristics of eight metals subjected to fully reversed cyclic strain from 0·2 to 50% were examined, the metals including low carbon and stainless steel specimens. Strain hardening characteristics peculiar to each metal were found, and the inverse relationship between the plastic strain range and the square root of the number of cycles to failure was verified in the case of the stainless steel over six decades of cycles of strain, ranging from $0{\cdot}25–2 \times 10^5$ cycles.

Determination of fatigue life by loading between variable limits of stress S. Luthander and G. Wallgren (*Aeronautical Res. Inst. Sweden, Report No.*18, 1946; *Translation* 421, *Royal Aircraft Establishment*, 1953, March, pp.19) The test method has special reference to aircraft and simulates gust loads.

An experimental study of the V-notch fatigue test D. S. Dugdale (*J. Mech. Phys. Solids*, 1959, **7**, Oct., 282–287) Rotating bending tests were carried out on sharply notched specimens of several materials. Logarithmic plotting gave S–N curves consisting of straight lines with a slope common to all materials, suggesting a power law of crack propagation. When residual stress was introduced by pre-loading, results following a similar pattern were obtained. Correlation between results of static and fatigue tests is discussed.

High-stress, low-cycle fatigue properties of notched alloy steel specimens S. Yukawa and J. G. McMullin (*ASTM Bull.*, 1959, (241), Oct., 38–40) Notched specimens from a large forging which failed in the notch test in <10000 cycles are described. Cases in which 10–20% overloads improved fatigue life are noted.

The effect of stopping holes on the development of fatigue cracks in plate G. Wallgren and E. Rudin (*FFA Rep. HU*–534, *MOS, TIL Ref. P*62813, *Royal Aircraft Establishment Trans.* 703, 1958, April, pp.16) Tests (on Al plates) are described.

Fatigue tests of plain plate specimens and transverse butt welds in mild steel R. P. Newman and T. R. Gurney (*Brit. Weld. J.*, 1959, **6**, Dec., 569–594) The authors describe the results of fatigue tests on a series of butt-welded specimens in mild-steel plate to BS 15. The effects of the method of edge preparation, of stress-relieving, and of the shape of weld reinforcement were considered, and the various fatigue strengths obtained have been related to that of the plain plate.—T.G.

Modifications of the crystalline structure produced by fatigue between 20 and 750 C on an ultra-mild carbon steel and on 18/8 austenitic steel J. de Fouquet (*J. Mech. Phys. Solids*, 1959, **7**, Oct., 264–271) [In French] Work was carried out on the changes of structure caused in an extra-low carbon steel and an 18/8 steel during alternating tension fatigue tests between 20 and 750°C. It was found that, below a certain temp. characteristic of the metal, the fatigue was shown by slip bands, and above it, by a progressive polygonization —and, in some cases, recrystallization in the course of the test.

Study of the initiation and development of fatigue cracks G. Wallgren (*Flygtekniska Forsoksanstalten Report No. HU*–569, 1955, Oct., *MOS. TIL Ref. P*62814; *Translation* 677 *Royal Aircraft Establishment*, 1957, July, pp.32) A report of a study under oscillating load of 0–8 kg/mm².

Study of the propagation of fatigue cracks G. Wallgren (*Flygtekniska Forsokranstelan Utosunda, Report No. HU*–471, 1955, Feb.; *MOS. TIL. Ref. P*62812; *Translation No.*700, *Royal Aircraft Establishment*, 1957, Oct., pp.49) Cr–Mo steel specimens were included. A pulsating load was applied to specimens with round holes and notches. The ratio of breaking load to area for different lengths of crack remained ~37 kg/mm².

The effects of sulphur on the notch toughness of heat-treated steels J. M. Hodge, R. H. Frazier, and F. W. Boulger (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 745–753) The investi-

gation described showed that impact energy levels on quenched and tempered alloy steel plates decreased continuously with increasing S content, relationship being linear on log coordinates. With cross rolling, the curves were raised for transverse and lowered for longitudinal tests. The basic analysis of the specimens was ~0·3%C, 0·8%Mn, 0·25%Si, 2·5%Ni, 0·8%Cr, and 0·45%Mn, and the S 0·005–0·179%.

Increase, by local heating, of the fatigue limit of plates and beams with welded-on gusset plates O. Puchner (*Soud. Techn. Connexes*, 1960, **14,** Jan.–Feb., 27–32) Pulsating bending fatigue tests on plates and beams with welded-on gusset plates and heat-treated to 700°C by local heating with an oxy-acetylene flame are discussed. The fatigue limit was found to be nearly the same as that of the base plate without gusset plates. It was not possible to determine with accuracy the influence of residual stresses due to local heating.—R.P.

Effect of the eutectic Pb–Sn melt on the endurance limit of carbon steel M. I. Chaevskii (*Metalloved. Term. Obra. Met.*, 1959, (8), 14–20) The experimental results of annealing 35 steel in a melt of the eutectic Pb–Sn alloy have shown an increased endurance limit of about 40% as compared with annealing in air.

The influence of the testing temperature on the fatigue strength of carbon steel specimens with stress raisers subjected to the action of the eutectic Pb–Sn alloy M. I. Chaevskii (*Fiz. Met. Metallored.*, 1959, **8,** (5), 789–791) Experimental findings described have shown that the fatigue limit of carbon steel specimens may be raised appreciably over a wide temp. range when the specimens are subjected to the action of the liquid eutectic Pb–Sn alloy.

The fatigue strength of sprayed copper coatings J. Salokangas and P. Lehto (*Eng. Dig.*, 1960, **21,** Jan., 105–106; from *Acta Polytechn. Scand.*, 1959, (6)).

Increased endurance strength of large cylindrical shafts with transversal openings M. P. Troitskaya (*Metallored. Obra. Met.*, 1959, (8), 12–14) Three types of steel with a UTS between 67·9 and 60·5 kg/mm² were examined under loads of 860–1 750 kg/cm² and the findings show the effectiveness of surface hardening as applied to cylindrical machine components with transverse openings. The endurance limit of specimens hardened by rollers and with high frequency current to a depth of 6·5–7·0 mm is greater by about one third than that of non-hardened specimens.

On the evaluation of creep tests V. Foldyna (*Hutn. Listy*, 1959, **14,** (12), 1172–1176) Various methods of extrapolating creep data, e.g. those of Larson and Miller, and of Glen, have been used with data obtained in extended creep tests on three low-alloy Cr–Mo–V steels containing also 0·07%Ni. Their respective reliabilities are discussed.—P.F.

Creep testing methods used in Czechoslovakia R. Pokorný (*Hutn. Listy*, 1959, **14,** (12), 1184–1187) Methods, machines, specimen shapes, etc. are discussed. Basically equipment and testing methods are similar to corresponding ones in use in most industrial countries.—P.F.

The use of curves correlating strain and strain rate in research on creep J. Glen (*Hutn. Listy*, 1959, **14,** (12), 1176–1180) Log. strain versus log. creep rate curves are shown to be particularly advantageous in the extrapolation of creep data. The probable theoretical significance of the shape of such curves is discussed and examples of their application are given, e.g. in the creep of Mn steels and low-alloy Mo–V–Ti steels.—P.F.

The relation between equilibrium diagrams and creep strength I. I. Kornilov (*Hutn. Listy*, 1959, **14,** (12), 1165–1168) It is shown that considerable information concerning the creep strengths of alloys can be derived from the equilibrium diagram. Reference is made to systematic development of high-creep strength in alloys in the USSR by the utilization of generalizations derivable from studies of the type of alloy structure indicated by the diagrams.—P.F.

Creep tests on a flake graphite cast iron at 400°C K. B. Palmer (*BCIRA J.*, 1959, **7,** Oct., 839–842) The limiting temp. in Great Britain, for the use of cast iron in steam engineering is 230°C; with a view to extend this limit to 340°C, creep tests were carried out at 400°C for 5000 h on unalloyed flake graphite cast irons of a type that had already been subjected to growth tests.—T.G.

On the extrapolation of results obtained in short-time creep tests to long times W. Siegfried (*Hutn. Listy*, 1959, **14,** (12), 1187–1193) Extrapolation methods, e.g. those of Larson and Miller, and of Glen are discussed. It is proposed that the log strain versus log strain-rate isotherms used by Glen be complemented by the use of the stress as the third variable, i.e. a three-dimensional creep surface is suggested as most suitable for purposes of extrapolation.

The function of boundaries and sub-boundaries in creep fracture of metals I. A. Oding, V. S. Ivanova, and J. P. Liberov (*Hutn. Listy*, 1959, **14,** (12), 1093–1095) The literature on the subject, and the authors own work on the development of creep fracture through heterogeneously nucleated void formation, are reviewed.—P.F.

The creep strength of some high temperature alloys at 1050–1200°C H. C. Child, A. B. Collier, and C. F. West (*Metallurgia*, 1960, **61,** Feb., 51–54) Tests on cast and wrought austenitic steels are described for periods up to 1000 h and strains up to 1%. Cast alloys showed some improvement and the best alloy examined was cast G·39 which had a creep strength at 1100°C, 1400 lb/in² greater than any other studied.

Microscopical observations on creep deformation of silicon steel at high temperatures T. Fujita and Y. Ishida (*Tetsu to Hagane*, 1959, **45,** Sept., 1119–1121).—K.E.J.

Effect of Mo, W, Ti, V and Nb on the creep-rupture strength of 12% chromium steel T. Fujita and T. Sasakura (*Tetsu to Hagane*, 1959, **45,** Sept., 1015–1017).—K.E.J.

High temperature creep and anelastic phenomena in polycrystalline refractory oxides R. Chang (*J. Nuclear Mat.*, 1959, **2,** July, 174–181) Creep in Al_2O_3 and BeO were studied and activation energies derived. The effects of small additions of Cr_2O_3 or La_2O_3 to Al_2O_3 and of MgO to BeO in improving high temp. ductility are discussed.

The effect of heat treatments on the mechanical properties and creep strength of the austenitic steel 15Cr35NiWTi L. Čížek (*Hutn. Listy*, 1959, **14,** (12), 1115–1118) Researches on the optimum heat treatment of the steel, which resembles closely Nimonic 80, are reported. The optimum solution treatment temp. is 1130°C, if best creep and fatigue strengths are to be obtained simultaneously.—P.F.

The damping capacity of heat-resisting steels and alloys, particularly under creep conditions J. Vodsedálek (*Hutn. Listy*, 1959, **14,** (12), 1125–1130) The effect of ageing on dislocation movement, and other structural effects due to heat treatments or associated with special features of the metallographic structure, were studied by means of a vibrator designed by the author. Steels examined included several high-alloy types, e.g. CrNiWV, CrNiVTi, CrNiW–MoTiAl, etc. Results are given for various heat treatments and the significance of the data is discussed.—P.F.

Stress relaxation in heat resisting austenitic steels R. Pech (*Hutn. Listy*, 1959, **14,** (12), 1119–1121) High temp. stress relaxation was studied in determination of the creep properties of two austenitic steels, one 18/8 CrNi(Ti) type and one 13/13CrNi steel containing 1·15%W, also Ti-stabilized. A simple automatic, self-registering relaxation machine is described.—P.F.

Some problems in the development of steels and alloys with a high creep strength M. V. Pridantsev (*Hutn. Listy*, 1959, **14,** (12), 1154–1158) Metallurgical problems arising in the development of steels for long-time service at high temp. are surveyed, with special reference to the author's own work and trends and researches in the USSR. High-alloy Cr–Ni steels, with or without the addition of other elements, such as Mo, W, Nb, and Ti are referred to in some detail.—P.F.

Experience gained in the development of creep resistant alloys N. P. Allen (*Hutn. Listy*, 1959, **14,** (12), 1161–1165) The basic principles which have guided research in the development of steels for use in steam turbines and for aeronautical purposes are surveyed, and the properties of some of the more important alloys are discussed.—P.F.

Study on AISI 321 type stainless steel for steel tubes used at elevated temperatures. III. Change of carbide of AISI 321 type stainless steel before and after creep rupture tests H. Hirano, S. Yamamoto, K. Yoshida, and H. Harima (*Tetsu to Hagane*, 1959, **45,** Sept., 1022–1025) The change in carbide composition is elucidated by electron diffraction and X-ray methods, and lattice constants are given.

Strength of rails with particular reference to rail joints S. Wise, D. Lindsay, and I. G. T. Duncan (*Chartered Mech. Eng.*, 1957, **6,** Dec., 486–487) A summary. Rail failures on British Railways are surveyed and fatigue tests on full sized rails are reported.

The 'Durimet' micro-hardness tester A. Metz (*Wire*, 1959, (42), Aug., 118) The use of the Leitz Durimet tester for measuring the hardness of wire is described.

Test blocks for indentation hardness testing J. G. Wood (*Metallurgia*, 1959, **60,** Sept., 115–118) Methods adopted at the N.P.L. for the production of standard hardness test blocks in the range 200–950 HV, for Vickers and Rockwell tests are described, and some results reported.

On the indentation of a ball. II. Strain figures developed in and on the ball indentation specimens S. Yazawa, K. Sato, M. Ueno, and G. Yaguchi (*J. Mech. Lab. Japan*, 1959, **5,** (1), 19–23) The object of this work was to determine the boundary of the plastically deformed zones in the specimens. Strain figures are given for various sections of specimens indented under different loads.

On the nature of the hardened layer in 'Indefinite chill rolls' R. Kamensky and J. Teindl (*Hutn. Listy*, 1959, **14,** (11), 971–977) Metallographic, fractographic, and other studies, together with analyses of carbides, were used to determine the nature of the hardening process occurring in the rolls, and to provide a sound basis for production technology.—P.F.

Increasing the accuracy of hardness measurements of metals B. M. Ovsyannikov and L. T. Timoshuk (*Zavods. Lab.*, 1959, (11), 1389–1391) Since the accuracy of hardness measurements is affected by the distance of the indentations from the edge of the test piece and by the distance between centres of two adjacent indentations, suggestions are made so that the All-Union Standards, and the International Organization of Standards ISO/P 79 be changed so that the minimum distance from the edge could be fixed at 2·5 d and that between two adjacent indentations at 4 d, in Brinell measurements.

On the edge-effect of residual stress. On the hardness distribution of a cold work roll body S. Onodera, Y. Arakida, T. Shibazaki, and K. Hori (*Tetsu to Hagane*, 1959, **45,** Sept., 1125–1127).—K.E.J.

Some findings regarding the change in the microhardness of technical iron heated within a wide range of temperatures, under tension in vacuo M. G. Lozinskii and V. S. Mirotvorskii (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 52–61) An apparatus is described where specimens can be heated up to 1300° and stressed in a vacuum of 10^{-5} mm. A series of tests were carried out on specimens kept at different temp. with ten indentations on each, with loads applied according to a definite pattern. Results show that the microhardness curves with tensile loads under isothermal conditions, have a certain minimum for certain definite loads with increasing temp. An interesting observation was that microhardness increases under stresses beyond these definite loads. This peculiarity was very pronounced for temp. up to 700°.

The mechanism of the softening of alloyed austenite with temperature M. E. Blanter, V. F. Gorshkova, and M. G. Lozinskii (*Metallored. Obra. Met.*, 1959, (7), 2–10) Experiments have shown that the softening of alloyed austenite with temp. obeys Shishokin's exponential equation and may be characterized by the initial hardness of the alloy H_0 and the temp. coefficient of softening α. The addition

of alloy elements brings about a change in the value of $H_{0(1)}$ of the initial hardness for the low temp. sections of the curve. Regarding the higher temp., additions of Cr, W, and Mo rapidly increase the values of the original hardness $H_{0(2)}$.

Method of determining hardenability and mechanical properties by quenching one end of the specimen Fan Chou Chun (*Acta Met. Sinica*, 1959, **4,** (3), 255–261) Flat specimens were used which were examined, after quenching, for structure and hardness, and after tempering, for mechanical properties corresponding to their acquired hardenability. A table was constructed from the data for hardness, structure, and distance from the quenched end, which may serve for the selection of suitable materials.

Factors influencing the hardenability of case-hardening steels A. Legat (*Hart. Techn. Mitt.*, 1959, **14,** (1), 51–67) The degree of homogeneity is stressed as an important factor in determining the hardening properties, and the limited applicability of end-quench and McQuaid–Ehn tests in assessing the effect of inhomogeneity on hardening properties.

Changes in the properties of mild steel caused by stress-ageing treatment. VII. The effect of stress on the process of strain-ageing of mild steel T. Kawasaki, H. Izumi, and T. Sakurai (*Nippon Kinzoku*, 1959, **23,** March, 141–144) Specimens of 0·05%C mild steel were strained up to 9% in tension and aged under stress between 55° and 250°C. Strain-ageing is accelerated by imposed stress. The max. increase of hardness and time to reach the max. value decrease with increase of stress. Activation energies show that the main process of strain-ageing under stress is the same as that of conventional strain-ageing.—K.E.J.

The temper brittleness of bainitic steels Tan Yu-hua (*Acta Met. Sin.*, 1959, Sept., **4,** 262–269) The tendency to temper-embrittlement in lower bainite is less than in tempered martensite; in upper bainite it is higher than in pearlite. That of tempered martensite is much more than that of pearlite. The temp. range is 460–580°C for upper bainite, and 430–620°C for tempered martensite. The temp. of max. embrittlement is 530°C for both, and the mechanisms are probably the same (25 refs).

Brittle-fracture strength of welded spherical container H. Kihara, K. Masubuchi, and H. Ishii (*Weld. J.*, 1959, **38,** Nov., 4518–456s) This work was concerned with the influence of residual stress, and the effects of stress-relieving heat-treatments, on brittle fracture. The containers were tested at various temp., and with and without notches. The results could be explained in terms of the theory advanced by Kihara and Masubuchi (*ibid.*, 1939, **38,** 159s–168s).

Hot-shortness of steel contacted with liquid metal M. Nakajima (*Tetsu to Hagane*, 1959, **45,** Sept., 1085–1087) Mechanical properties of steel samples heated in argon and dipped into various (Cu + Pb) liquid alloys are given, and microstructures of the (Cu + Pb) phase are described.—K.E.J.

Correlation between notch ductility and brittle fracture V. Komárek (*Hutn. Listy*, 1959, **14,** (9), 753–758) In specimens with a 3 mm notch, a linear relationship was found in 0·6%C steels between relative notch impact strength and brittle fracture, and a high degree of correlation was found in the derived regression equation (11 refs).

The problem of brittle fracture of mild steel in welded ship construction S. K. Basu (*Trans. Inst. Marine Eng.*, 1959, **71,** Nov., supp. vi–xii) The factors producing brittleness, especially the triaxial stress systems brought about by notches and service and welding stresses. Transition temp. and the factors affecting it are then reviewed. Suggested specifications for satisfactory shipbuilding steel are then drawn up.

The effect of impurities on embrittlement of low-alloy steels J. Mennen (*Schweissen Schneiden*, 1960, **12,** April, 156–157) The two types of embrittlement occurring in low-alloy steels are the 350°C and the temper embrittlement. The studies were conducted so that high-purity steels were tested and the steels were then treated with impurities and tested again. For the 350°C embrittlement it was found that P, Sb, As, Sn, Si, Mn, and N may cause brittleness. Temper brittleness was caused by P, even in amounts of only 0·008% and by As, Sn, Mn, and Si.—T.G.

Hydrogen brittleness of the 20 and 20 Kh steels with electrolytic saturation K. V. Popov and V. A. Yagunova (*Metalloved. Term. Obra. Met.*, 1959, (8), 51–52) These annealed steels were investigated after quenching from 950° and tempering at 600°. The electrolytic specimens were saturated with H_2 in N H_2SO_4 (with addition of sodium arsenate as activator; 3 mg As/l). The findings show that appreciable reduction of plasticity of the low carbon steels occurs even with a H_2 content not exceeding 3–5 ml/100 g. of metal (0·0002–0·0004% by weight).

Wear of metals (*Mét. Const. Méc.*, 1960, **92,** March, 171–183) The mechanics and causes of wear, with observations on the wear of copper and its alloys, on metal-to-metal friction, the effects of surface oxides, and various processes for improving resistance to wear are reviewed.

How wear is influenced by surface finish A. Sonntag (*Met. Prog.*, 1960, **77,** April, 140–142) Abstract of a paper 'The significance of surface finish on friction, wear and lubrication' by the above author, presented at Lucerne, Sept. 1959. The influence both of amount and geometry of surface irregularities on lubrication are discussed.

A theory of the abrasion of solids such as metals J. Goddard, H. J. Harker, and H. Wilman (*Nature*, 1959, **184,** Aug. 1, 333–335) Studies of emery paper of different grades on various metals are reported. Values of the coefficient of friction are obtained and interpreted.

An experimental study on the abrasion of martensitic steel weld metal S. Ito and M. Kawaguchi (*J. Mech. Lab. Japan*, 1959, **5,** (1), 52–58) The work described was on the abrasion resistance of hard-facing weld metal in various states (e.g. sorbitic, martensitic).

Temperature effect of the second quenching of case-hardened steels on their impact wear resistance S. S. Ermakov and V. I. Kochnev (*Metalloved. Term. Obra. Met.*, 1959, (8), 49–50) The method of testing impact wear resistance was outlined in *Zavods. Lab.*, 1957, (9), by Vyaznikov and Ermakov. An increased temp. for the second quenching results in an increase of residual austenite and in a coarsening of the martensite needles. Such structural changes have been observed in 20KhN3A and 20N3MA steels. The appearance of residual austenite and its uniform distribution increases the strength and plasticity of the cemented layer and also the impact wear resistance. When quenching from very high temp. the volume of residual austenite increases and also the martensite needles and this results in a reduction in the strength of the cementite layer.

These machines, grind, scratch, gouge to measure abrasive wear D. Gardner (*Prod. Eng.*, 1959, **30,** Nov. 9, 70–73) An illustrated review.

Wear of carbon steel weld metal. II. The effect of previous heat treatment of the specimens upon the wear and the structurally altered layers of the worn surface S. Ito, K. Honda, and K. Ishiyama (*J. Mech. Lab., Japan*, 1959, **5,** (1), 24–29) This work describes the effect of previous heat treatment upon the wear resistance of a specimen, and the effect of previous heat treatment on the layers of the steel that are changed structurally by the wearing process. **III. The effect of the rubbing conditions upon the wear and the structurally changed layers of the worn surface** S. Ito and K. Ishiyama (30–34) The effect of rubbing conditions on the wear of carbon steel weld metal, and the structure of the surface layers after wear, has been investigated. It was found that, whether the sliding distance, the sliding velocity, or the load were varied, the effect was the same. For this reason it is thought that only the temp. of the surface layer is of importance.

Wear of Ni–Cr austenitic steel weld metal. I. A study on the wear and the structurally changed layer of the worn surface S. Ito and K. Honda (*J. Mech. Lab., Japan*, 1959, **5,** (1), 44–51) The wear of this steel due to dry metal-to-metal contact was investigated, and analysed in terms of rate of wear, and the structural changes in the surface layer.

Erosion resistance of case-hardened steel and its sub-microstructure I. M. Lyubarskii, A. P. Lybuchenko, and G. N. Bakakin (*Fiz. Met. Metallored.*, 1959, **8,** (6), 872–877) The most favourable structure of steel from the point of view of resistance to erosion may be determined by observing correctly the processes taking place in the active layer during friction and is governed by the kinetic hardening or softening of the different structural components during friction. This resistance is affected also by the phase composition and submicrostructure of the phases. In general, therefore, when phase composition and microstructure change, the resistance to erosion depends on the content of austenite. An appreciable proportion of residual austenite imparts good erosion resistance during friction of the hardened layer in which the γ-phase crystals are in a non-hardened condition while those of the α-phase are work-hardened.

Spectral emittance of ceramic-coated and uncoated specimens of Incomel and stainless steel J. C. Richmond and J. E. Stewart (*J. Amer. Ceram. Soc.*, 1959, **42,** Dec., 633–640) Specimens coated in six ways were examined at 900, 1200, 1500, and 1800°F over the range 1·5–15·0 μ.

Absolute method of measuring magnetic susceptibility A. Thorpe and F. E. Senftle (*Rev. Sci. Instr.*, 1959, **30,** Nov., 1006–1008).

Neutron study of the crystal and magnetic structures of $MnFe_2$–Cr_2O_4 S. J. Pickart and R. Nathans (*Phys. Rev.*, 1959, **116,** Oct. 15, 317–322).

Anomalies of reactivity at the Curie point in iron vanadium alloys M. Daire (*Compt. Rend.*, 1959, **249,** Nov. 16, 2065–2067) Oxidation is affected by the amount of ferromagnetic constituent and by the mode of association of the two components.

The observation of domain structure in magnetic thin films by means of the Kerr magneto-optic effect M. Prutton (*Phil. Mag.*, 1959, **4,** Sept., 1063–1067) Blooming the polished surface and using a quarter wave plate enables the effect to be used to observe domain structure. Results on 81%Ni–Fe in a steady field are reported.

Calculation of the hysteresis loop of an Alnico permanent magnet alloy with preferred crystallisation W. Baran (*Techn. Mitt. Krupp*, 1959, **17,** (3), 150–152) The hysteresis loop was calculated for Koerzit-700 alloy, using the Stoner and Wohlfarth models. The results agree well with the experimental values, if the magnetic saturation and volume ratios assumed by Tenzer and Kronenberg and the known electron microscopical structure are taken as a basis.

The investigation of mechanical and magnetic properties of the iron–nickel–aluminium alloy for permanent magnets A. A. Shekalov (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 88–98) Addition of 0·2%Si to a Fe–Ni–Al alloy produces the best combination of magnetic and mechanical properties. Ti additions may be used in alloys with Co by works producing alloys for different combinations of coercive force and residual induction. Best testing methods of these steels is under torsion.

Temperature dependence of the magnetostriction constants in iron and silicon iron E. Tatsumoto and T. Okamoto (*J. Phys. Soc. Japan*, 1959, **14,** Nov., 1588–1594) The magnetostriction constants λ_{100} and λ_{111} of Fe, 1·08% and 3·83%Si–Fe have been measured by using single crystals at temp. below the Curie temp. and down to −186°C. Present theories do not appear to give an adequate explanation of the experimental results.

Texture and magnetisation curves of iron–silicon sheet with cubic texture H. Stäblein (*Techn. Mitt. Krupp*, 1959, **17,** (3), 159–161) Type and distribution of primary and secondary recrystallization texture in thin cubic-texture sheets are described.

Changes in the saturation magnetization and the electrical resistance of iron–nickel alloys on uniform compression at low temperatures E. I. Kondorskii and V. L. Sedov (*Zhur. Eksptl. Teoret. Fiz.*, 1958, **35,** (4), 845–853).

Paramagnetic studies of dilute hydrogenated

palladium–iron alloys J. P. Burger, E. Vogt, and J. Wucher (*Compt. Rend.*, 1959, **249,** Oct. 19, 1480–1482).

Variations in the factors g and g′ in Fe–Ni alloys G. Asch (*Compt. Rend.*, 1959, **249,** Oct. 19, 1483–1485).

Transitions from ferromagnetism to antiferromagnetism in iron–aluminium alloys. I. Experimental results A. Arrott and H. Sato (*Phys. Rev.*, 1959, **114,** June 15, 1420–1426). **Theoretical interpretation** H. Sato and A. Arrott (1427–1440).

New electron diffraction apparatus for continuous recording R. Thun (*Rev. Sci. Instr.*, 1959, **30,** June, 399–407) Fe_3C formation was one reaction studied with the apparatus described. Only an outline of applications is given.

Radiographic sensitivity data for the isotopes cobalt 60, iridium 192, caesium 137, thulium 170 and thorium 228 M. B. Anderson (*Nondestr. Testing*, 1959, **17,** Nov.–Dec., 365–370) Curves of exposure/thickness for 2% sensitivity were obtained on a fine-grained medium and a fast film for slotted iron wedges up to 9 in. thick, and also Al.

Coulomb excitation of high energy nuclear levels in even isotopes of tungsten D. G. Alkhazov, A. P. Grinberg, G. M. Gusinskii, K. I. Erokhina, and I. Kh. Lemberg (*Zhur. Ekspt. Teoret. Fiz.*, 1958, **35,** (6), 1325–1334) States of ^{182}W, ^{184}W, and ^{186}W were identified.

X-Ray study of the solid solution of TiO_2, Fe_2O_3, and Cr_2O_3 in mullite (3 Al_2O_3, 2 SiO_2) M. Krishna Murthy and F. A. Hummel (*J. Am. Ceram. Soc.*, 1960, **43,** May, 267–273).

The K X-ray absorption spectrum of iron at the temperature of liquid nitrogen S. A. Nemnonov and K. M. Kolobova (*Fiz. Met. Metallov.*, 1959, **8,** (3), 478–480) The results of the analysis at + 20 and −180° are compared in a table.

Critical potentials of soft X-rays emitted from iron T. Hayasi and H. Kawaharada (*Sci. Rep. Tohoku Univ.*, 1959, **43,** June, 1–5).

An electron diffraction study of the effects of heat treatment on α-Fe_2O_3 (haematite) single crystals M. Blackman and G. Kaye (*Proc. Phys. Soc.*, 1960, **75,** March, 364–368).

Radioisotopes in metallurgy W. A. Mudge (*Met. Prog.*, 1960, **77,** March, 101, **169A,** 170) A brief outline report of a conference on the subject held at Philadelphia in Jan. 1960.

Elastic scattering of protons by vanadium, chromium, iron and cobalt C. A. Preskitt and W. P. Alford (*Phys. Rev.*, 1959, **115,** July 15, 389–397).

Elastic scattering of 3·7-Mev neutrons from 8, Fe, Co, Ni, Cu, and Zn M. K. Machwe, D. W. Kent jun., and S. C. Snowdon (*Phys. Rev.*, 1959, **114,** June 15, 1563–1570).

Properties of malleable iron at elevated temperatures L. C. Marshall, G. F. Sommer, and D. A. Pearson (*Met. Prog.*, 1960, **77,** March, 102–107) High-temp. mechanical properties of ferritic and pearlitic malleable cast iron are reported.

High strength stainless steel for high temperature service (*Metallurgia*, 1960, **61,** Jan., 25–26) An account of the Jethete series of low-carbon alloy steels.

Adiabatic calorimeter for metals in the range 50–1 000 C E. E. Stansbury, D. L. McElroy, M. L. Picklesimer, G. E. Elder, and R. E. Pawel (*Rev. Sci. Instr.*, 1959, **30,** Feb., 121–126).

New iron-base high-temperature alloy? G. K. Manning (*Met. Prog.*, 1960, **77,** April, 117–119) The properties are described of several 20%Cr alloys with Ni, Nb, and C, some of which show very high hardness at elevated temp. The preliminary work so far carried out is described.

Chromium steels remain tough after eight years at high temperature M. A. Scheil (*Met. Prog.*, 1960, **77,** April, 104–112) Long-time tests on specimens of various type 405 stainless steel specimens placed inside oil refinery equipment indicated that compositions at the lower end of the Cr range (< 13%Cr) and with small additions of Mo possess good resistance to embrittlement at 715° and 900°F.

Heat resistance of the iron–chromium–aluminium alloy No.2 between 600–1200° V. S. Mikheev (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 66–72) The alloy, in work-hardened condition, annealed below 700°, i.e. recrystallized with a fine grain of 0·001–0·002 mm² deforms more rapidly than when annealed between 850–1000° with 0·009–0·05 mm grains. With increasing annealing temp. and grain size the resistance of the alloy under test under an applied load increases. In bending tests the heat resistance of the alloy in 3–5 mm thick specimens for 500 h was found to be: at 600°, 4 kg; 700°, 1 kg; 900°, 0·3 kg; 1000°, 0·2 kg, and at 1 200° for 300 h, 0·025 kg/mm².

The effect of cold hardening on the heat resistance of the austenitic heat-resistant steels Chou Shi Chan and M. L. Bernshtein (*Acta. Met. Sinica*, 1959, **4,** (3), 235–244) The steels examined were EI 395 and EI 69. Cold-hardening appreciably increases their heat-resistance. The special features of their fine submicroscopic nonuniformity is the principal cause. It has been shown that recrystallization exerts a considerable effect on their heat resistance properties only when the rate of growth of the recrystallizing grains is very high, and then the effect of cold hardening is rapidly eliminated.

How low temperatures affect nine high strength alloys R. L. McGee, J. E. Campbell, R. L. Carlson, and G. K. Manning (*Mat. Design Eng.*, 1959, **50,** Nov., 106–107) 17–7 PH, AM–350, 4340, and INCO 300–M steels are tabulated at temp. down to −423 F.

Investigating the mechanical properties of solid bodies, especially metals at 4·2 K and below. III. Mechanical properties of iron, titanium, tantalum, and St.2 steel at 4·2 K and lower O. V. Klyavin and A. V. Stepanov (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 922–927) The method of investigation was fully explained (*Fizika Tverdogo Tela*, 1959, **1,** (6)). The results of the present investigations on polycrystalline specimens of these metals at 300, 78, 4·2, and 1·6°K have been tabulated under various headings such as UTS, yield point, impact strength, reduction of area, elongation, etc.

Diffusion processes in steels containing arsenic D. S. Kazarnovskii and T. M. Ravitskaya (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 15–27) In high carbon steels, a high arsenic content increases microliquation to a large extent between the branches of the dendrites. As shows a large diffusivity into the austenite of a high-carbon steel which explains fully the disappearance of the banded structure after a diffusion anneal at 1 100° for 8 h; When heated to a high temp. in an oxidizing medium a pronounced enrichment with As occurs in a thin metal layer under the scale which results in the formation of a bright film containing much As.

Grinder for sectioning solid diffusion specimens H. W. Schamp jun., D. A. Oakes, and N. M. Reed (*Rev. Sci. Inst.*, 1959, **30,** Nov., 1028–1031) The device is intended for use with radioactive tracers.

Diffusion of some alloying elements in liquid iron T. Saito, Y. Kawai, K. Maruya, and M. Maki (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1959, **A, 11,** Oct., 401–410) [In English] The diffusion coeff. of Si, S, and Mn in liquid iron were measured to be in the range $2–5 \times 10^{-5}$ cm²/sec. at low concn., and the activation energies 6–10 kcal/mole. Up to 20%, the coeff. increased with Si content in pure and carbon-saturated liquid iron. The coeff. of Si and S decreased in presence of C (10 refs).—K.E.J.

The effect of vanadium additions on the self-diffusion of iron M. S. Zelinskii, B. M. Noskov, P. V. Pavlov, and E. V. Shitova (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 725–730) The self-diffusion of iron in Fe–V alloys with different V and S contents was investigated by means of radioactive iron. The volume and boundary diffusion parameters have been established and also their change with the change in the concentration of the second element. The activation energy of the self-diffusion of iron decreases linearly with increasing V content. The effect of V in iron contrasts with that of the transition elements (Cr, Mn, Ni) which increase the activation energy of self-diffusion.

Diffusion of ^{63}Ni in iron, cobalt, nickel, and two iron–nickel alloys J. R. MacEwan, J. U. MacEwan, and L. Yaffe (*Can. J. Chem.*, 1959, **37,** Oct., 1629–1636) Self-diffusion of Ni and diffusion of ^{63}Ni into Fe, Co, and two Fe alloys containing 5·79 and 14·9%Ni respectively was studied by measuring decrease in surface activity. Ni was found to diffuse more slowly in Fe than in the Fe-rich portion of the Fe–Ni system, and the rate of Ni diffusion increased with Ni content.

Overvoltage and diffusion through iron and palladium J. N. Andrews and A. R. Ubbelohde (*Proc. Roy. Soc.*, 1959, **253,** Nov. 17, 6–15) X-Ray diffraction studies of cathodes during discharge and diffusion of H_2 are described. Diffusion in Fe occurs through grains rather than along boundaries.

Low alloy steels Yu. A. Geller and Ts. L. Olesova (*Metalloved. Obra. Met.*, 1959, (7), 30–45) The technological, and a number of other properties of carbon tool steels, can be improved by alloying them with small additions of Cr, W, and V, thus obtaining a better uniformity in the distribution of the excess carbides and a smaller tendency to develop a cementite lattice. These additions also preserve a much smaller grain-size in heating and give lesser sensitivity to overheating, a strength of the steels in the quenched and low tempered conditions greater by 10–20% for hypereutectoid and by 20–30% for eutectoid and hypoeutectoid steels; better quenching in hot media, which reduces deformation appreciably, greater hardenability and a better distribution of the hardness in tools with diameters up to 30–35 mm, are also achieved.

The physical metallurgy of precipitation-hardenable stainless steels D. C. Ludwigson and A. M. Hall (*DMIC Rept.* 111, *AD* 214001; *PB* 151067, 1959, April 20, pp.70; from *US Res. Rep.*, 1959, **32,** Dec., 11 751) Martensitic, semiaustenitic, and austenitic types are reviewed. Handling, heat treatment, and fabrication are also considered.

Investigating high speed steels of great efficiency A. P. Gulyaev and S. M. Saverina (*Metalloved. Obra. Met.*, 1959, (7), 22–30) Eight new types of steel were investigated for which optimum conditions for quenching and tempering have been worked out. The nonuniformity of the carbides in small cross-sections does not depend on the composition, but in large cross-sections it is determined by the W content. Steels with 18%W show rather large carbide liquation. Steels with higher Co, V, and C contents when quenched from the optimum temp. have a higher proportion of residual austenite than the R18 steel. However, a threefold tempering of 1 h each time at 580° ensures its almost complete transformation.

Stainless 17% chromium steel (*Aciers Fins. Spéc.*, 1960, March, 19–23) The physical and mechanical properties of a 17%Cr, 0·07%C steel are discussed in relation to its use. This steel is not usually heat treated except for annealing. It can be stamped, drawn, and welded by several methods and shows good corrosion resistance.—R.P.

Comparisons of materials (*Mat. Design Eng. Materials, Selector Issue*, 1959–60, 1959, **50,** 11–31) A series of tables including steels giving specific gravity (12) thermal conductivity (13), thermal expansion (14), modulus of elasticity (19), tensile strength (20–21), elongation (22–23), hardness 24–25), impact strength (26), yield strength (27), creep strength (28), stress-rupture strength (29), and effects of radiation (30–31). **Properties of materials: irons and steels** (37–72) Physical and mechanical properties are tabulated.

New steel alloy J. H. Schaum (*Mod. Castings*, 1959, **36,** (Sept., 43) A new series of hard, impact- and erosion-resisting alloy steels, developed by Eimco Corp., Salt Lake City, can be welded without embrittlement in the heat-affected zone, and have high corrosion resistance. They contain 0·8–1·3%C, up to 4%Cu and 2·5%Mn, 5–10%Ni, 3% max. Cr and Mo, and varying amounts of Si, V, Ti, Co, and W.

Fundamental study on Mo V low-alloy heat-resisting steel K. Beguchi and H. Nakajima (*Tetsu to Hagane*, 1959, **45,** Sept., 1115–1117) Data are given for room- and high-temp. mechanical properties, optical and electron micrographic examinations, transformation points and S-curves.—K.E.J.

On the precipitation-hardening stainless steels F. Tsukamoto (*Tetsu to Hagane*, 1959, **45,** Nov., 1282–1298) A review (38 refs).

Studies on alloy A286. I. On the mechanical properties of alloy A286 T. Hasegawa, O. Ochiai, and S. Yamashita (*Tetsu to Hagane*, 1959, **45,** Sept., 1029–1031) Tensile and creep-rupture properties of the vacuum- and air-melted material are given, and compared with those of Timken 16–25–6.—K.E.J.

301 Stainless modified for 800 F and up D. B. Roach, A. M. Hall, R. Sergesen, and A. R. Stargardter (*Space/Aeronautics*, 1958, Oct., pp.2, reprint) An account of MicroMach cold-rolled steel, a 301 grade containing N_2.

On the characteristics of 'Cons-el' arc melted bearing steels Y. Yagi and M. Kurihara (*Tetsu to Hagane*, 1959, **45,** Sept., 992–995) The steels are compared with air-melted steels in respect of composition, gas analysis, inclusions, microstructure, hardness, strength, and surface defects.—K.E.J.

The activities of the Verein Deutscher Eisenhüttenleute during 1959 (*Stahl Eisen*, 1960, **80,** April 28, 581–615) Among the activities of the Verein reported here are those of the publication department and the various committees on special subjects. The work done by the Max Planck Institut für Eisenforschung is included in this report and so is that on standardization and specifications. More than 300 references are given.—T.G.

METALLOGRAPHY

New horizons in metallography R. W. K. Honeycombe (*Metallurgist*, 1960, **1,** Jan., 34–41) A brief account of recent replica techniques for the electron microscope, and their results.

Contribution to the metallurgical and metallographic study of cast steel P. Détrez and M. Goguillon (*Fonderie*, 1960, Jan., 11–15) A study of transformation temp. and of the tendency to austenite to exhibit grain growth is made with a view to recommending heat treatment cycles for cast steel. The dilatometer is advocated for determination of transformation points and a vacuum method is described for determining austenitic grain size.—R.P.

Metallic compounds I. I. Kornilov and B. K. Vul'f (*Uspekhi Khim.*, 1959, **28,** (9), 1086–1113) After a short historical introduction, the article gives a succinct review of the state of knowledge on this subject including a classification of the physical and chemical origins of metallic compounds; the conditions for their formation in solid solutions, general properties, characteristics of the different classes of compounds, i.e. Kurnakov's compounds (e.g. Fe_3Pt–$FePt$–$FePt_3$) compounds with a valency ratio between atoms (Mg_2Sn, Mg_3Sb_2, etc.), electron bonds, compounds in which components A and B are in the proportions AB_2 such as $MgCu_2$, $MgNi_2$, etc. compounds of the type of nickel arsenides such as $NiTe_2$ etc. hydrides, borides, carbides, silicides, nitrides, etc. (159 refs).

Metal structures can be analysed simply, inexpensively L. H. Burbage (*Canad. Metalw.*, 1960, **23,** April, 39–41) A metallograph for the photomicrography of cast iron is described. This is American Optical Metalstar microscope and 682G camera with Polaroid (Land camera) back.

Principles of high-temperature microscopy B. B. Brenden, H. W. Newkirk, and J. L. Bates (*J. Am. Ceram. Soc.*, 1960, **43,** May, 246–251) A literature summary of objectives, stages, and methods of heating (48 refs).

Two-stage replica method for electron microscopy J. L. Miles, H. F. Stedman, and R. M. Cornish (*Rev. Sci. Instr.*, 1959, **30,** Aug., 746) Metal films after vacuum deposition are examined by depositing a thick layer of SiO (3000 Å) and then an epoxy resin with amine catalyst. The SiO layer can then be used to make further replicas.

Techniques for the direct examination of metals by transmission in the electron microscope P. M. Kelly and J. Nutting (*J. Inst. Met.*, 1959, **87,** Aug., 385–391) Methods of preparing foils are classified and assessed, and factors in satisfactory electrolytic preparations are considered.

An outline of the theory of diffraction contrast observed at dislocations and other defects in thin crystals examined by transmission electron microscopy M. J. Whelan (*J. Inst. Met.*, 1959, **87,** Aug., 392–405).

Filament furnace for microscopy studies at high temperatures H. W. Newkirk jun. and J. L. Bates (*Rev. Sci. Instr.*, 1959, **30,** Aug., 645–646) A device for heating 1 g samples to ~3000° is described. Vacuum or controlled atmosphere can be used.

Second-stage graphitization in a nodular cast iron J. Burke (*JISI*, 1960, **196,** Sept., 42–45) [This issue].

Investigating the pearlite transformation in isolated austenite powder A. P. Gulyaev and V. D. Zelenova (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 475–476) The method of obtaining a powder of isolated austenite from quenched steel by anode solution was described in the *Zavodska Lab.*, 1959, (1), 60. The pearlite transformation was studied on powders of Kh12F1 steel (1·49%C; 12·0%Cr; 0·28%V) precipitated by electrolytic solution. The diagrams of the isothermal decay of austenite in the domain of the pearlitic transformation of the isolated and the non-isolated austenite correspond. It was therefore considered that it is similar to the pearlite transformation in the compact steel specimen.

Investigating the intermediate transformation of the isolated austenite powder A. P. Gulyaev and V. D. Zelenova (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 476–478) The findings show that, as in the case of the martensite transformation, the intermediate transformation in the isolated austenite prepared by electrolytic solution is absent, while it takes place normally in the compact specimen.

Diagrams for the transformation of austenite in Kh6VF tool steel (*Supplement to Metalloved. Obra. Met.*, 1959, (7), 31) This steel is intended for cold metal stamping, for cutting tools on lathes for light work and for other machine components requiring great hardness, resistance to wear, small deformability on quenching, and a hardenability to penetrate right through. This steel is used for the same purposes as Kh12 steel but shows less liquation and has better mechanical properties. It has however a slightly lower wear resistance. Its composition is 1·0–1·15%C; 5·5–7·0%Cr; 1·1–1·5%W; 0·5–0·7%V; 0·45%Mn; 0·35%Si; 0·03%S; 0·03%P; 0·35%Mo.

The effect of the austenitizing temperature and the supercooling of the interrupted anneal on the amount of residual austenite in alloy steels E. S. Tovpenets and F. L. Ivanov (*Metalloved. Obra. Met.*, 1959 (7), 50–57) In the interrupted anneal with supercooling to 100–400° and then heating from the supercooled temp. to 650°, the residual austenite is maintained; this is shown by the increased strength. The amount and the strength of this residual austenite depends not only on the chemical composition but also on the austenitization and on supercooling since the quantity of residual austenite increases with increasing austenitizing and supercooling temp.

Transformation of austenite during the continuous cooling of some structural steels used for components of large cross-section M. P. Braun, B. B. Vinokur, and F. I. Ivanov (*Metalloved. Obra. Met.*, 1959, (7), 58–60) The possibility of constructing thermo-kinetic diagrams by dilatometry has been demonstrated. These diagrams constructed from curves for the isothermal transformation of austenite are not sufficiently exact but may be used in the absence of data obtained directly. The kinetics of the austenite transformation in structural steels with W or Mo and with similar values for the other alloy elements are the same in continuous cooling.

Application of the effect of low stresses to the transformation kinetics of austenite E. Zmihorski (*Draht*, 1960, **11,** April, 170–172) The principle and practice of a straightening process are described in which parts such as tools are transferred from the salt bath, in which they have attained the austenitic condition, into a device in which they can be stressed. Stress has to be applied before reaching the martensite transformation temp. as this temp. increases when a stress is applied.

Isothermal decomposition of the austenite in ARC 3 Steel B. Rotenstein and M. Dragan (*Studii Cercet. Metalurg.*, 1959, (4), 493–510) The investigation has shown that over the pearlitic range the kinetics of transformation show the normal features of the isothermal reactions for this temp. The intermediate range shows two distinct zones, an upper between 450–350° in which transformation forms upper bainite and pearlite, and a lower zone in which lower bainite is formed. A critical point for the curve of the beginning of the transformation was found at about 340° which shows increased resistance to the transformation of the supercooled austenite at this temp. This point separates two zones between 325 and 350° in which the transformation, after prolonged holding, does not extend to more than 60% of the initial volume of the austenite.

Process of carbon precipitation from martensite and the formation of carbide in the tempering of chromium- and chromium-tungsten alloy tool steels with about 1% carbon H. Kramer, E. Kramer, K. Ebner, and M. Storek (*Hütt. Techn. Mitt.*, 1959, **14,** (1), 24–50) It is shown that C-precipitation from martensite is retarded by alloying additions; this effect is very noticeable in the case of Cr and W. The mechanism is discussed. In the steels investigated—100 Cr4, 100 Cr 6, and 105 W Cr 6 only unalloyed carbide was precipitated on tempering up to 400°C. In the Cr–W steel Cr-alloyed cementite and the cubic carbide of type $Fe_{21}Me_2C_6$ also separated at higher tempering temp In the first stage of tempering ε-carbide was also apparently formed in the low-alloy steels, and carbide grain growth appears to be influenced by the ratio C: alloying metal (34 refs).

The martensite transformation in thin films of iron alloys W. Pitsch (*J. Inst. Met.*, 1959, **87,** Aug., 444–458) Electron microscopical studies are reported.

The effect of composition on the structure and properties of martensite K. J. Irvine *et al.* (*JISI*, 1960, **196,** Sept., 66–81) [This issue].

Stabilization of the martensitic transformation in iron–nickel alloys J. Woodilla, P. G. Winchell, and M. Cohen (*Trans. Met. Soc. AIME*, 1959, **215,** Oct., 849–851) A study of the kinetics of stabilization with respect to the isothermal component of the martensite reaction in Fe–Ni alloys show that C and N contents, although low, play an important part in stabilization, of which no evidence is found in their absence. The fact that, in the presence of 0·007%C, activation energy of the stabilization process is comparable to that for the diffusion of C or N in ferrite rather than in austenite suggests that the interstitial diffusion controlling the stabilization occurs in the martensite embryos rather than in the parent phase.

Study of the iron carbon diagram R. Hagége and M. Génot (*Compt. Rend.*, 1959, **249,** Nov.23, 2195–2197) Heats of transformation of pearlite and heat of solution of graphite in austenite are discussed.

Intermetallic phases in two types of modified 12%Cr steels J. Pluhář, H. Tůma, J. Ježek, and R. Pospíšil (*Hutn. Listy*, 1959, **14,** (12), 1108–1112) Laves phases occurring in 12%Cr steels destined for work at 600–620°C were studied. The steels were modified with Mo and V, or with W, Co, and V.—P.F.

The types of defects in alpha-leboite F. A. Sidorenko, P. V. Gel'd, and L. B. Dubrovskaya (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 465–466) Experiments here described were carried out on alloys prepared in W vacuum furnaces from chemically pure Si (99·95%) and R–4 carbonyl iron. The alloys were homogenized at 1080° for 100 h in vacuo. The results show that the ζα phase (α-leboite) is stable within the limits of 53·5 and 56·5%Si, which confirms the findings of Haughton and Becker (*JISI*, (1930), **121,** 315) but contradicts the conclusions of Phragmén (*JISI*, 1926, **114,** 397). **The nature of the ζα-phase in the Fe Si system** ((5), 735–739) Specimens of iron and silicon alloys containing from 50 to 58%Si by weight were analysed by X-rays. The structural features of the α-leboite may be explained by the formation of solid solutions either by substitution or by replacement. Either model agrees with a reduction in the parameters of the ζα-phase lattice with increasing Si concentration. The experimental data here explained point however to the fact that the high temp. ζα-phase consists of a solid solution.

CORROSION

Translations of foreign terms on corrosion of metals (*Corrosion*, 1960, **16,** Jan., 32t–36t) A table is presented in which terms commonly used in corrosion control work are given in German, English, French, and Russian. This table has been reprinted from a German DIN publication; in an Appendix the meaning of those terms which differ from that of the table as understood in the United States is given.

The theory of pitting and action of a centrifugal field on the pitting corrosion of 18–8 chromium nickel steels W. R. Fischer (*Techn. Mitt. Krupp*, 1959, **17,** (3), 137–145) Experiments in which unstabilized 18–8 steel was subjected to the action of a chloride-containing solution in a centrifugal field support the theory that increased halogen ion concentration exists at the active, anodic locations of pitting and that the halogen ions take part directly in the dissolution of metal.

Effects of diffusion on corrosion of metals by fused salts C. Edeleanu *et al.* (*JISI*, 1960, **196,** Sept., 59–61) [This issue].

Some observations on the interaction of liquid sodium with cast irons and plain carbon steels A. A. Smith and G. C. Smith (*JISI*, 1960, **196,** Sept., 29–41) [This issue].

ANALYSIS

Determining chromium in chromium steels from the value of the e.m.f. P. D. Korzh (*Zavods. Lab.*, 1959, (11), 1342–1344) An account is given of a comparison between measurements obtained from the thermo-emf and the tribo-electric method. In the first case the thermo-emf can be expressed as a linear function of the Cr content so that the Cr content can be obtained from the readings for emf. The tribo-electric force, occurring during friction between two conductors, is determined by the chemical composition of the two friction specimens and their structure and to a lesser degree by the friction. The results of the experiments agree well with tests on standards of Armco iron.

Amperometric determination of molybdenum in steel with solid electrodes S. K. Churkov and L. S. Studenskaya (*Zavods. Lab.*, 1959, (9), 1034–1036) The method is based on the precipitation of Mo^{VI} and Mo^{V} salts with lead salts in an acetic acid medium. The indicator for the electrode reaction is the reduction of Pb^{2+} at the platinum cathode. The special feature of this method is the use of two similar platinum electrodes instead of a calomel auxiliary electrode.

Determination of molybdenum in alloys by means of quinoline-8-thiol R. B. Golubtsova (*Zhur. Anal. Khim.*, 1959, **14,** (4), 493–496) The method is applied to Ni- and Ti-base alloys containing Cr and other elements, with a modification in the presence of W. The effect of other metals, including Fe^{3+}, is also considered. A small sample weight (0·01 g) is used, and the determination is photometric.

Complexometric determination of molybdenum after its reduction to the quinquevalent state A. I. Busev and Chang Fan (*Zhur. Anal. Khim.*, 1959, **14,** (4), 445–450) Mo^{5+} is titrated with excess 0·01 *M* EDTA, and the excess titrated with 0·01 *M* $ZnSO_4$, with Chromagen black ET as indicator, giving a sharp end-point colour change from bright green to reddish-brown. A modification is suggested for direct photometric titration of Mo^{5+} with EDTA.

The determination of tantalum and niobium in ores and ferroalloys. 2 H. Wirtz and H. Rothmann (*Z. Erz. Met.*, 1959, **12,** Dec., 612–613) Ta and Nb are isolated by an Na_2O_2 fusion followed by hydrolysis, and determined as described in an earlier report.

Separation of uranium from iron and vanadium K. Singh and D. Patnaik (*Proc. Indian Acad. Sci.*, 1959, **50A,** Dec., 358–363) Photolysis in sunlight in NH_4F, HF solutions containing alcohol is described. U precipitates as a complex fluoride.

Differential thermoanalysis J. A. Schedling (*Radex Rundsch.*, 1959, (4), 600–611) Principles and applications of the technique are discussed (31 refs).

The spectrophotometric detection of end-points J. Smeddon (*BCIRA J. Res. Dev.*, 1959, **7,** Dec., 885–894) An account is given of experiments carried out from time to time in the BCIRA Scottish laboratories. Photometric titration has been used mainly for determination of Mg in nodular cast iron; Ca in ferro-silicon, and MgO in limestone and slag.

Quantitative spectrochemical analysis of carbon steels (rimming) for silicon A. Kh. Gavryutina and M. V. Uzlyuk (*Zavods. Lab.*, 1959, (12), 1478) Standard apparatus is used and the arrangement and the method of preparing the specimens is described. It takes 15 min. to photograph the spectrum of a specimen twice.

A new tool for case carbon evaluation A. Goldblatt (*Met. Prog.*, 1959, **76,** Sept., 87–88) C gradient in steel is measured by a direct-reading vacuum spectrograph (the Quantovac). The counter-electrode was a pointed Ag rod, and depth of penetration during the exposure was 0·001 in. Close agreement was obtained with the results of chemical analysis.

Spectrochemical determination of aluminium in large concentration in a Fe Al alloy N. V. Buyanov and N. P. Sukhova (*Zavods. Lab.*, 1959, (12), 1479) This method determines Al in quantities between 10 and 30% and uses forged specimens as a source of the arc with a carbon electrode.

Spectrophotometric determination of silicon in ferrous metals R. Boulin (*Compt.-rendu XXXIst Congrès Internat. Chim. Industrielle*, 1958, Sept., Liège; preprint, pp.5) The application of heteropolyacids to the determination of Si in metals is reviewed, and the conditions studied for the formation of the blue Si–Mo complex, and for the elimination of the effects of P, As, V, Cr, W, Ni, Co, Mo, and Fe.

Spectrographic analysis of thin metallic films L. E. Berenshtein and V. G. Koritskii (*Zavods. Lab.*, 1959, (11), 1344–1345) Condensate forming on copper laminate probes introduced into the chamber containing metal vapours are the specimens for this analysis. The film thickness obtained is 1–2 microns. Various systems have been analysed: Fe–Cr, Fe–Mn, Ni–Cr, Ni–Co, and the analytical spectrum lines are supplied for each system. A comparison between chemical and spectrographic analysis shows deviation of 0·96–3·5%.

Spectrographic method of determining iron impurities in titanium turnings V. V. Gusarskii and I. A. Kuzovlev (*Zavods. Lab.*, 1959, (12), 1464–1465) According to data in the relevant literature, mass and area of the surface subjected to radiation affect the blackening of the analytical lines of the various components. The experiments here described were conducted to verify this finding. An arrangement is described in which the turnings are placed in the lower electrode in which a small depression has been bored for the purpose. The pair of analytical lines selected is Fe 2599·40 Å and Ti 2555·99 Å. The results show that mass affects individual results according to the base of the alloy and to the element which is being determined.

Spectrographic analysis of non-ferrous metal slags N. L. Fishkova and M. M. Rysina (*Zavods. Lab.*, 1959, (12), 1459–1461) An account is given of the analysis of powdered slag introduced into the discharge zone of the spectrograph. A subsequent chemical analysis for comparison purposes shows differences of ±3·5 to 6·0%.

Complete spectrochemical analysis of fireclay and chamotte N. V. Semkina and A. B. Shaevich (*Zavods. Lab.*, 1959, (12), 1465–1467) A method is described for a full and simultaneous analysis of SiO_2, Al_2O_3, TiO_2, CaO, and MgO and separately of Na_2O and K_2O. A comparison between the spectrographic and chemical analysis of different materials shows differences of the order of ±1%.

A source of pulses for analysing slags from sintered burdens M. N. Shtutman, V. P. Avdeenko, E. V. Zvereva, V. M. Shteinberg, and T. A. Eremeeva (*Zavods. Lab.*, 1959, (12), 1457–1459) A scheme for using the sparking pulses from a PS–39 generator is described, because, with increasing distance between electrodes the darkening by the analytical lines rapidly loses intensity the discharge is confined by a disc. Experiments with such discs are described. Variations in height from 2 to 8 mm and in dia. of the opening from 1·5 to 7 mm do not greatly affect the darkening of the lines. With a small disc (2–4 mm) the material of the specimen is scattered in the shape of relatively large particles of fused metal.

Modern methods of slag analysis W. E. Clarke (*BICRA J. Res. Dev.*, 1959, **7,** Dec., 867–884) Disadvantages of classical and abbreviated methods for analysis of cupola slags are explained; a method is described in detail which enables a complete and accurate analysis for seven major constituents to be carried out in 3 to 4 h. One sample is dissolved and dehydrated with $HClO_4$ to yield the SiO_2. A second sample is dissolved and fumed with $HClO_4$ to volatilize the SiO_2; six portions of the remaining solution are analysed as follows: FeO, TiO_2 by spectrophotometric methods; Al_2O_3, CaO, MgO by volumetric methods; MnO by either technique. Alternatively the FeO can be determined volumetrically on a third sample.—D.L.C.P.

Colorimetric determination of boron in porcelain enamel frits J. J. McIntosh and J. E. Cox (*J. Amer. Ceram. Soc.*, 1960, **43,** March, 123–124) A method using curcumin is given.

Chemical analysis of a calcined kyanite from Kenya, E. Africa H. R. Shell and G. W. Martin (*US Bur. Mines Rep. Invest.*, 5557, 1959, pp.9) Seven laboratories determined ten constituents and the reasons for discrepancies are discussed.

Rapid analysis of ore and similar substances A. Neuberger, E. Schöffmann, and K. Herkenhoff (*Arch. Eisenh.*, 1960, **31,** Feb., '91–96) Directions are given for the exact estimation of P, TiO_2, Cu, and Co in ore and similar substances by the borax fusion method. Further directions are included for the estimation SiO_2, Ni, V, and As.—T.G.

INDUSTRIAL USES AND APPLICATIONS

Present tendencies in engineering and metal structural work M. R. Delesques (*Mém. Soc. Ing. Civils France*, 1960, March, 15–23). **Qualities of steel and its capacities** M. J. Bernard (24–28) Weldability is discussed, also qualities of structural steels and the use of tubing. **Conclusions** M. P. Lorin (29–30).

High grade reinforcing steel for reinforced concrete construction P. Mlosch (*Neue Hütte*, 1960, **5,** Feb., 70–78) An account is given of the historical development of reinforced concrete; types of reinforcing steel are enumerated and their properties described, the development of prestressed concrete is outlined, and some requirements of reinforcing steel are reviewed (19 refs).

Chain: Pick and buy it wisely H. F. Reid, jun. (*Iron Age*, 1959, **184,** Nov. 26, 102–103) A brief description is given of the properties and recommended applications of chains made from wrought iron, low carbon steel, high carbon steel (heat treated), and low alloy steel (heat treated). A cost comparison is also made.—A.G.

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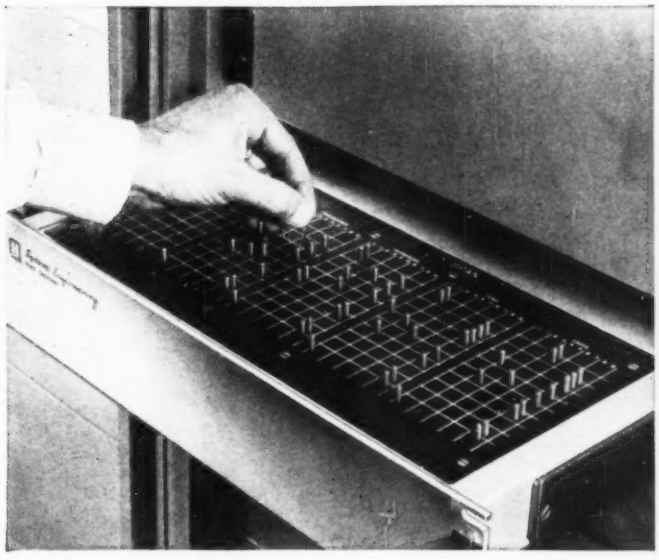

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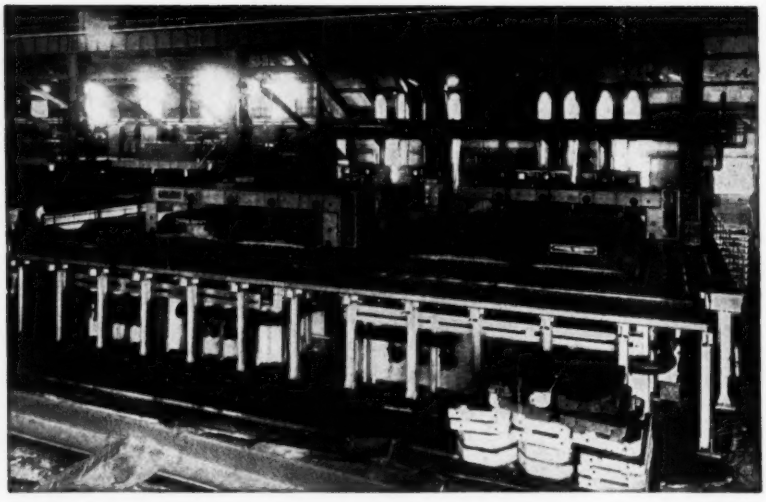

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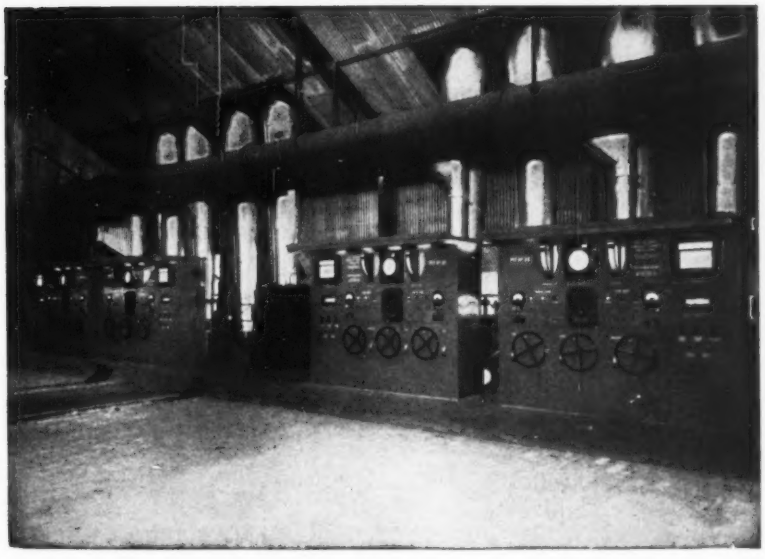

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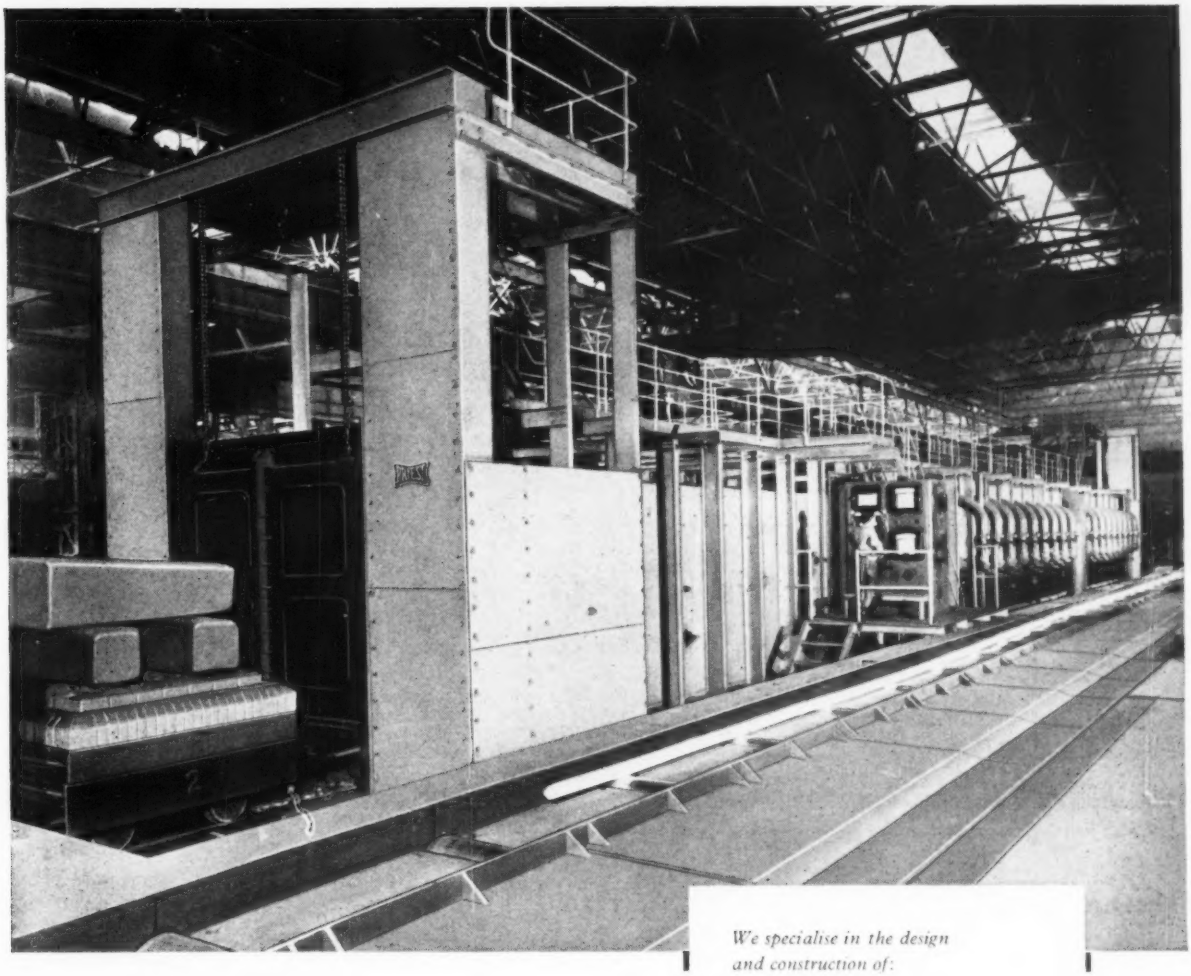

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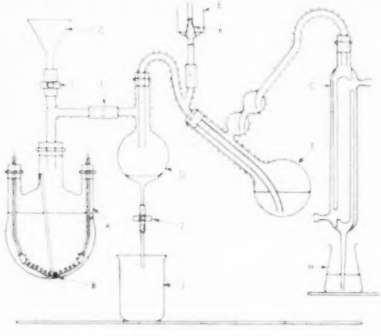

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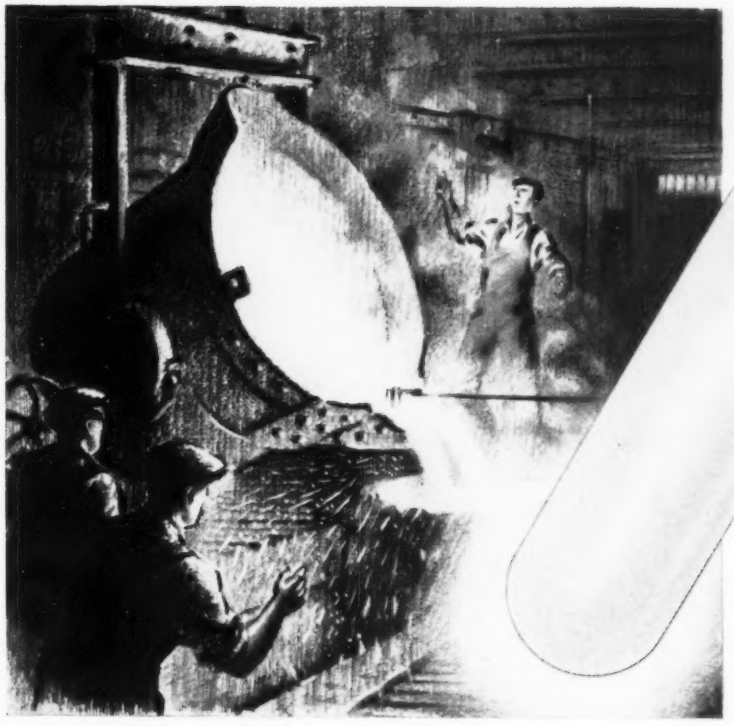

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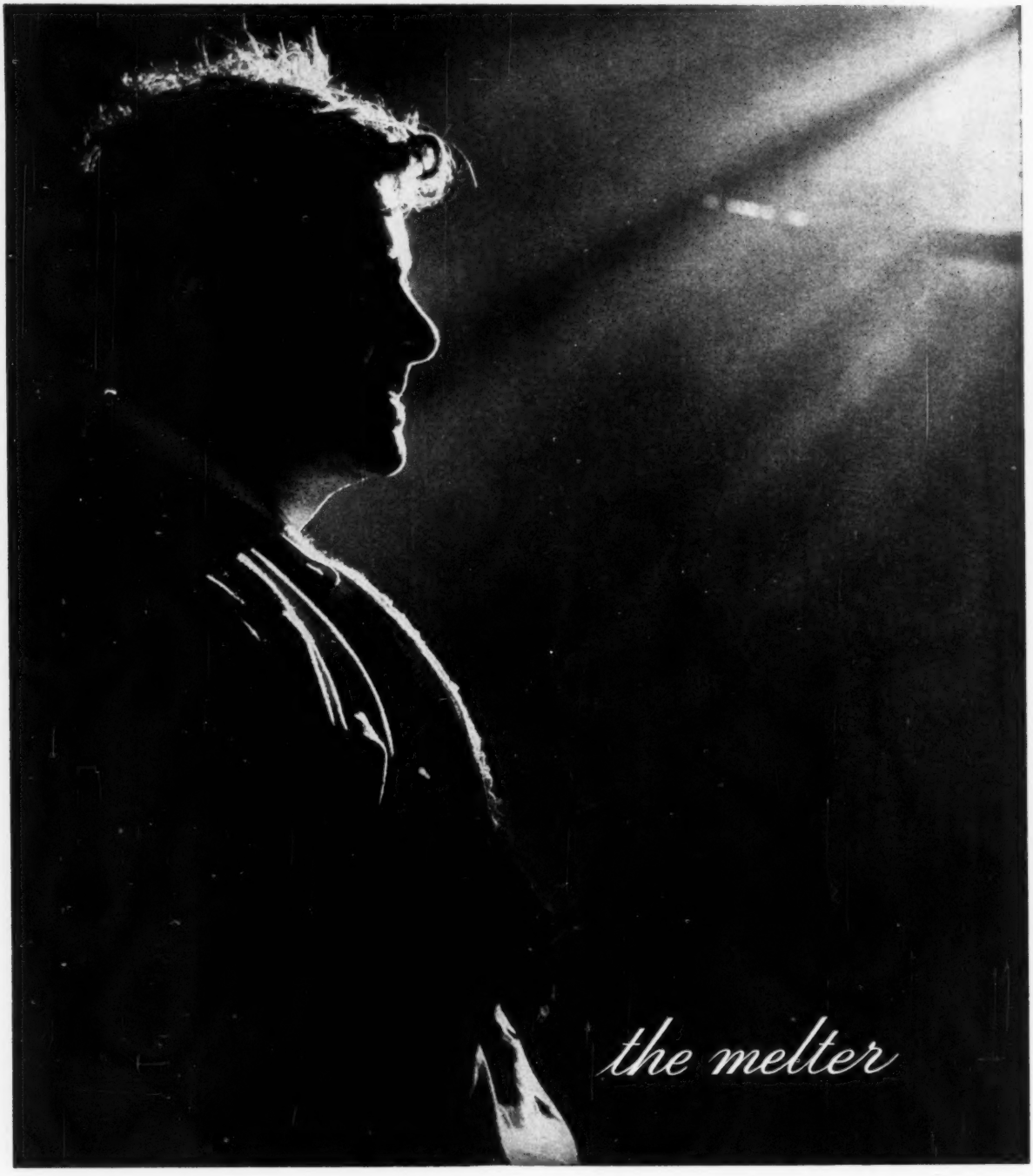

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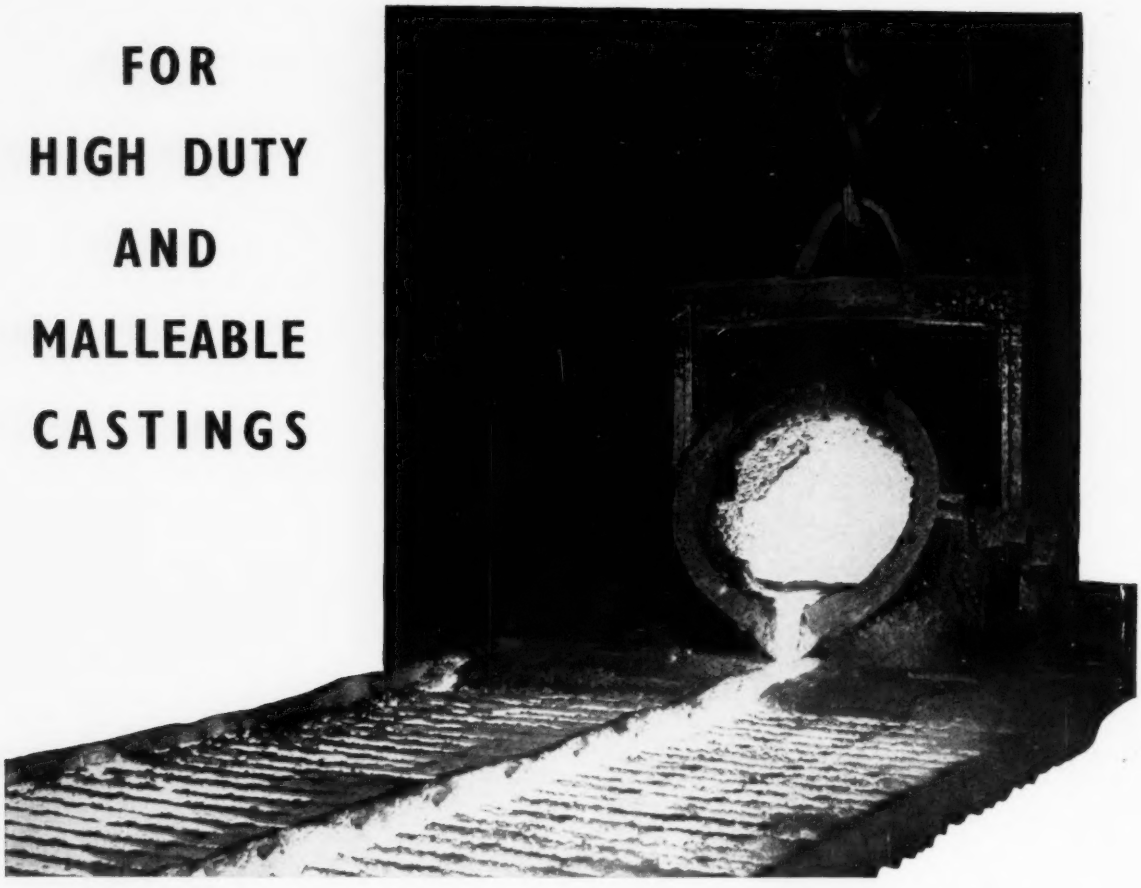

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PHOSPHORUS%	0·6–0·7	0·6–0·7	0·6–0·7	0·4–0·5	0·25	0·25	0·25
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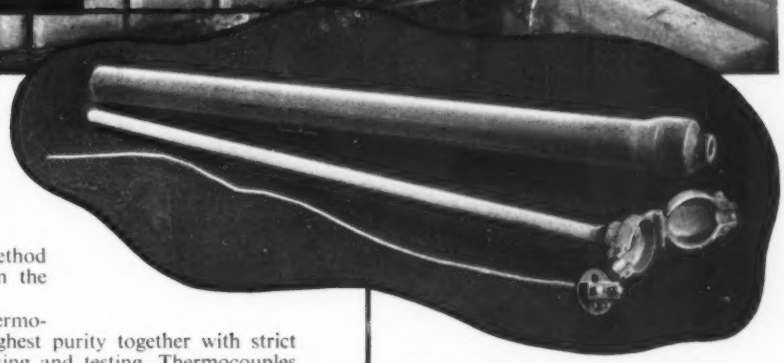

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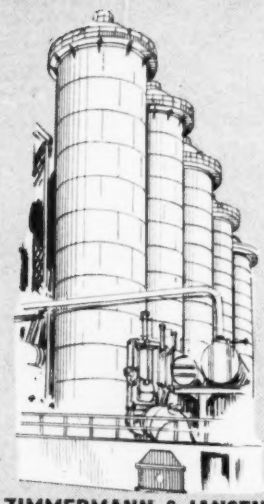

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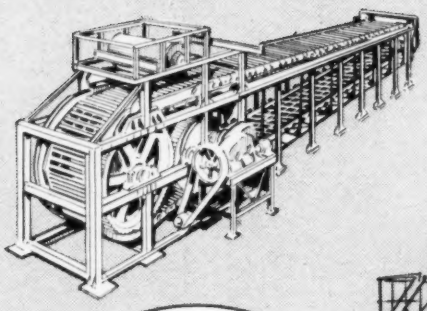

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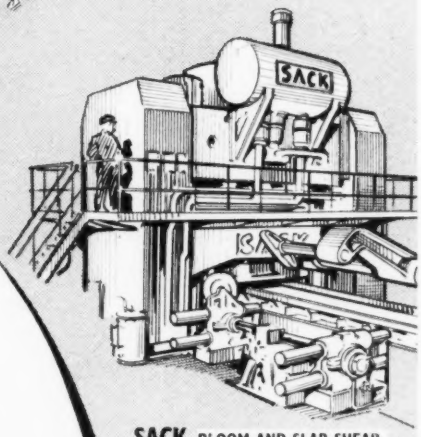

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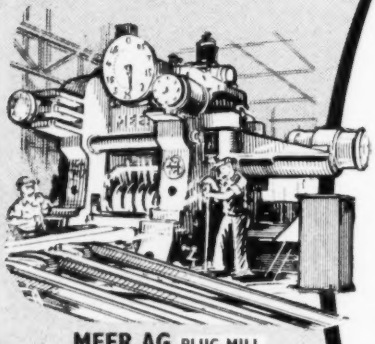

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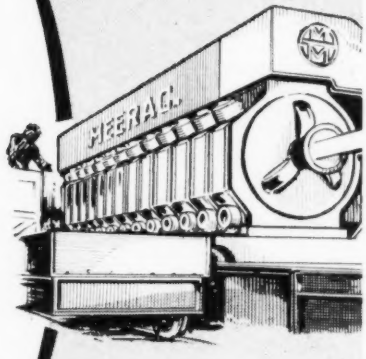

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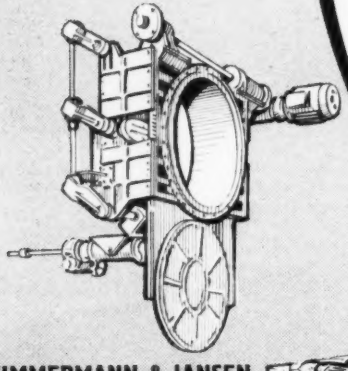

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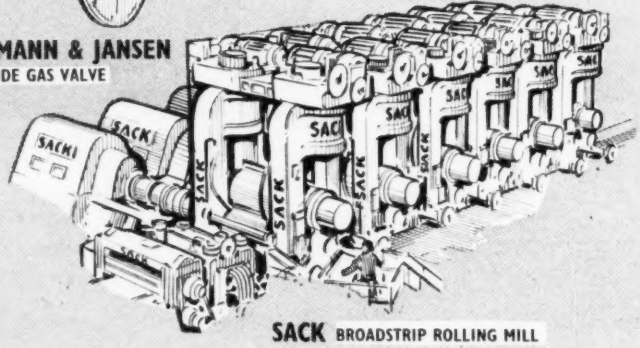

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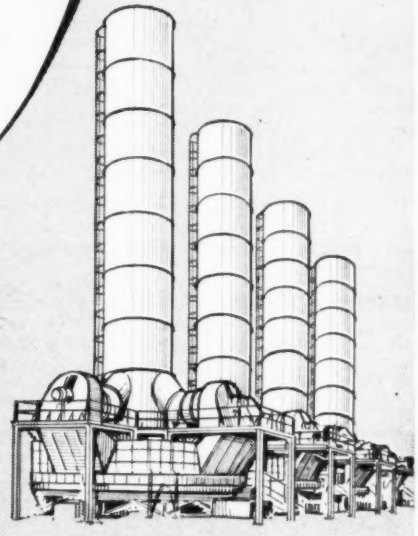

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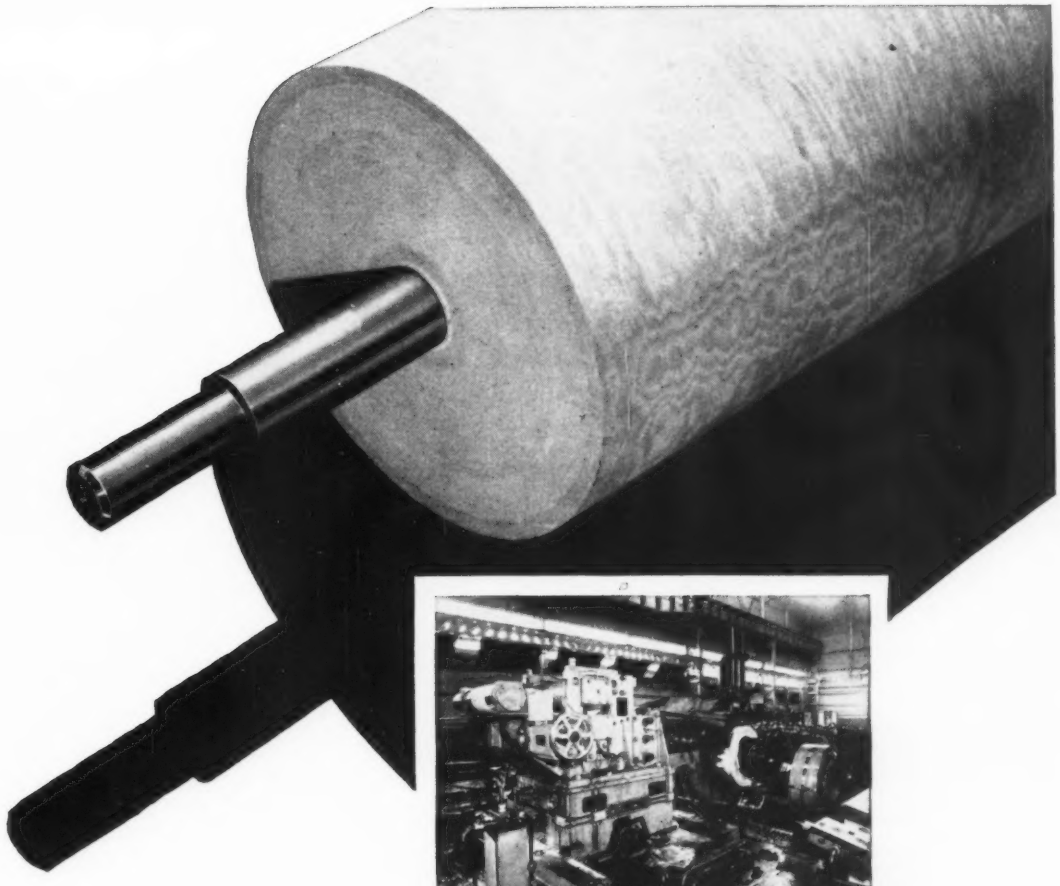

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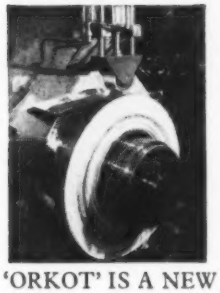

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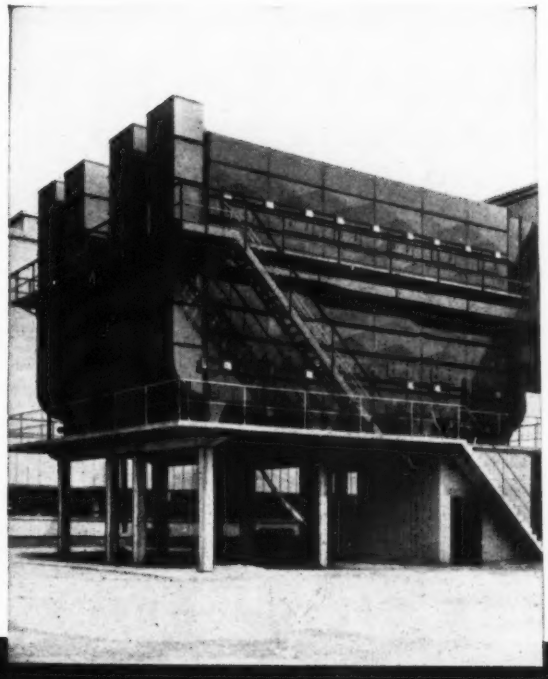

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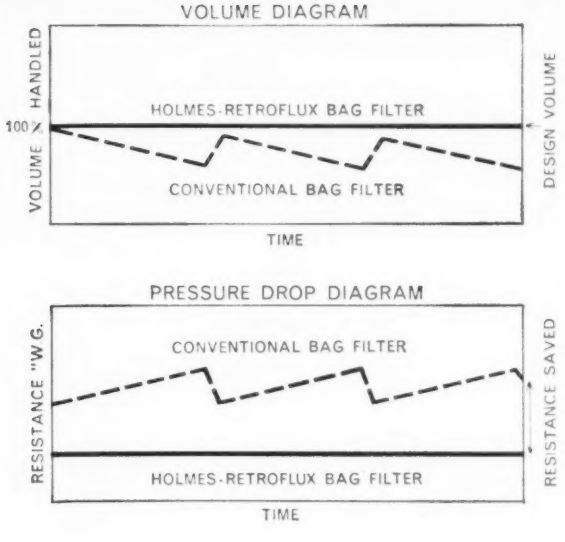

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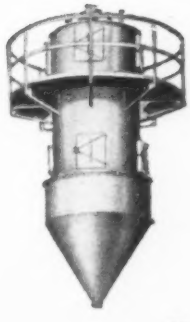

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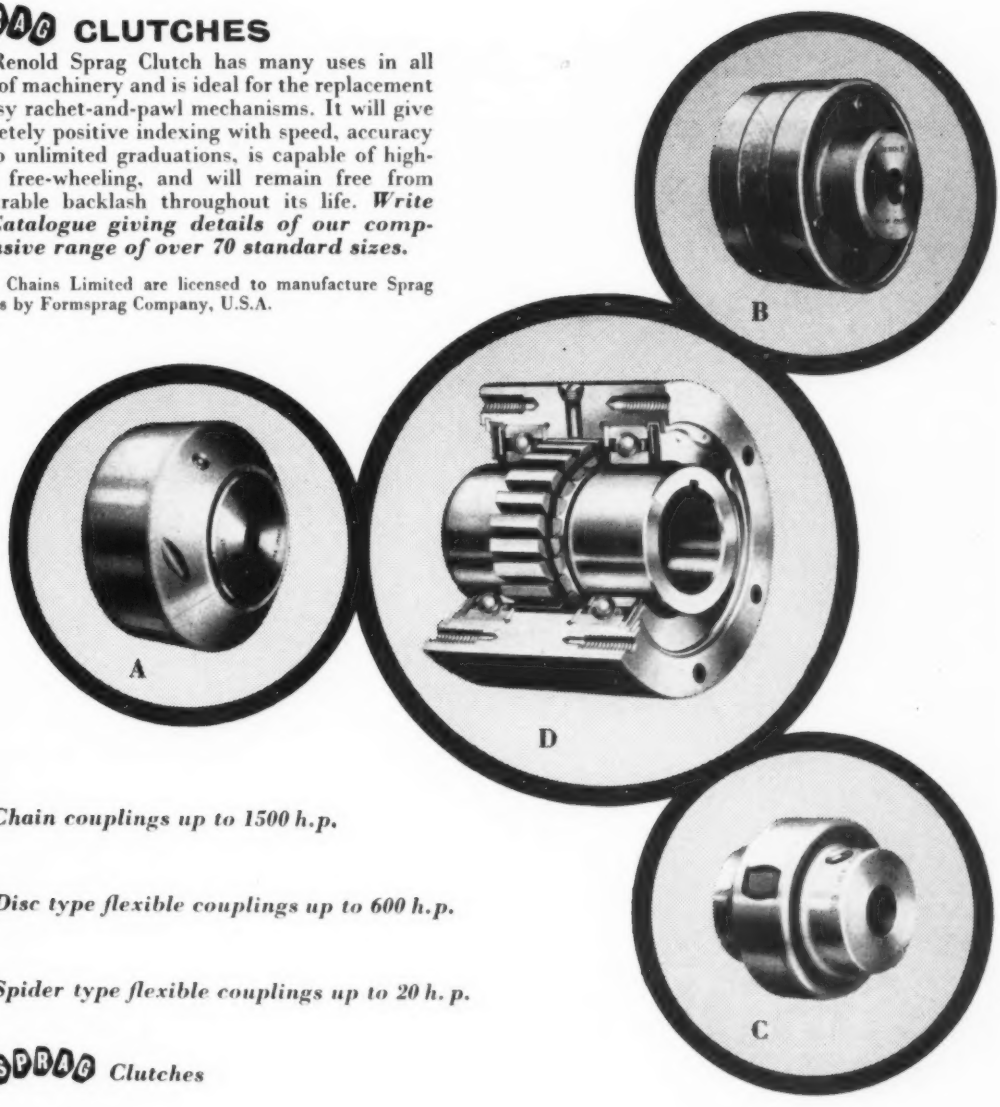

A *Chain couplings up to 1500 h.p.*

B *Disc type flexible couplings up to 600 h.p.*

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D SPRAG *Clutches*

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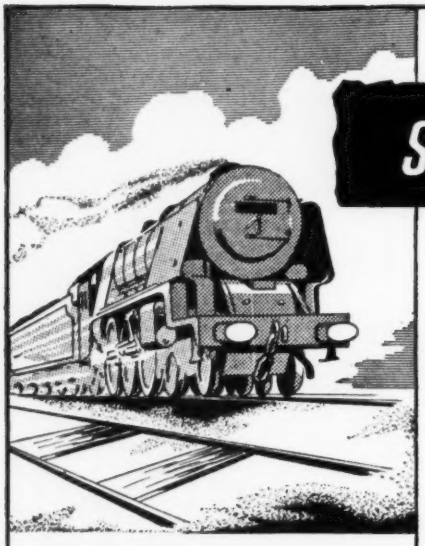

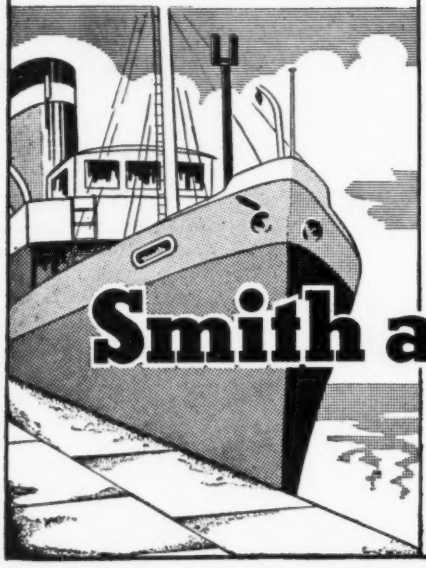

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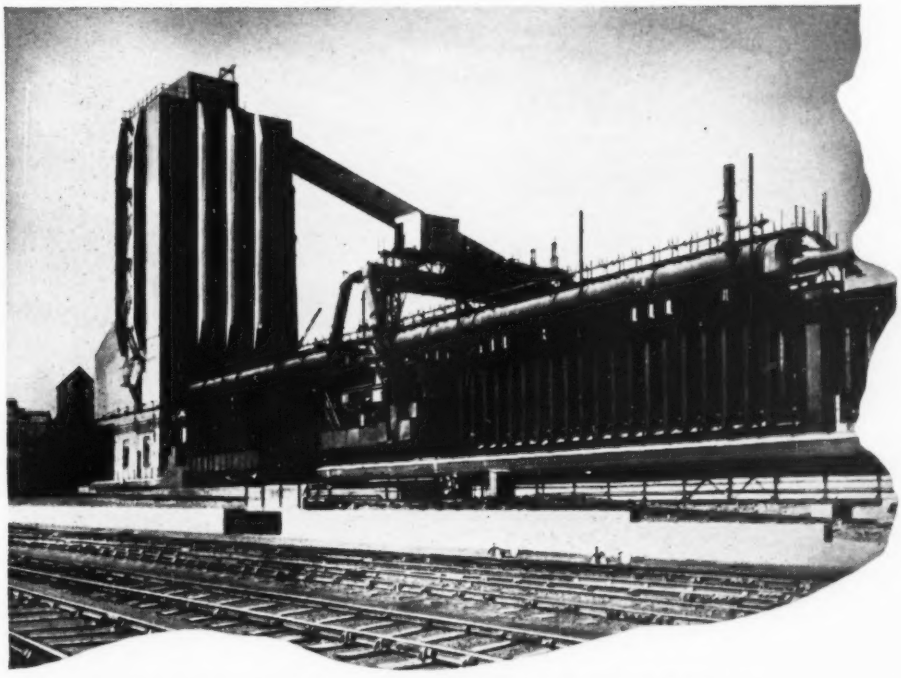

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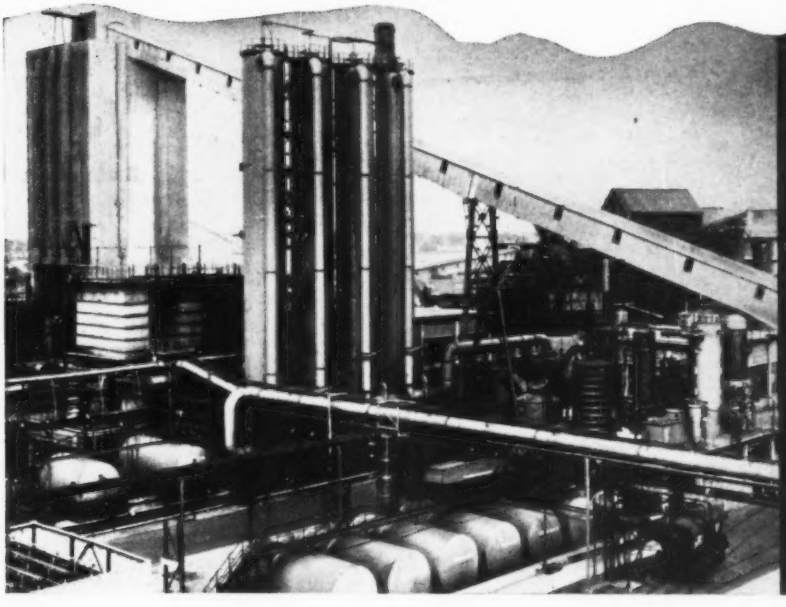

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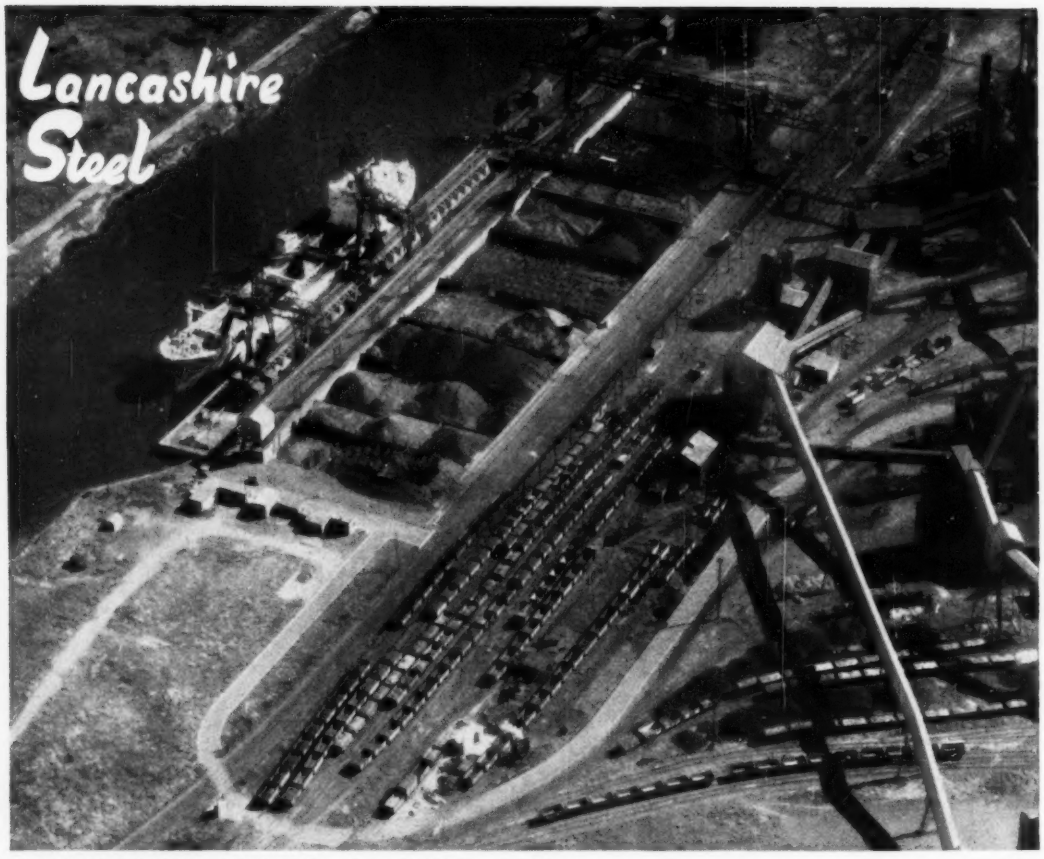

AERIAL VIEW OF THE UNLOADING WHARF—*Photograph by Aero Pictorial Ltd.*

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Dielectric Heating - 3

Some further details of the uses to which dielectric heating can be put are given in this data sheet, being continued from data sheet No. 11.

The Woodworking Industry

A most important development in recent years in the woodworking industry has been the introduction of synthetic resin adhesives of the thermosetting type for the bonding and adhesion of wooden components.

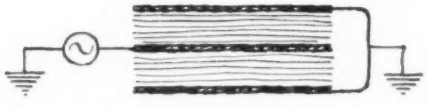
Setting of these resin adhesives proceeds at a rate largely determined by temperature. For instance, urea formaldehyde, one of the resins in common use, sets as follows:

TEMPERATURE	SETTING TIME
65 F	3 hours
80 F	1 hour
150 F	3 minutes
190 F	1 minute

The resultant bonded joint is equally satisfactory in each case. Most of the power supplied when dielectric heating is used is absorbed by the resin, the heat thus being concentrated where required and power consumption reduced to a minimum.

Plywood

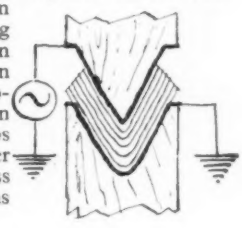
With dielectric heating consuming power only during the heating cycle, plywood can be produced with considerable savings in heating times and costs.



For example, in a press holding 100 3-ply $\frac{1}{2}$" thick assemblies, the resin glue is set in 20 to 30 minutes, depending upon the dryness of the wood. An output of up to 60 cu. ft. of plywood is obtained per hour using a 25 kW H.F. generator.

Curved Laminated Sections

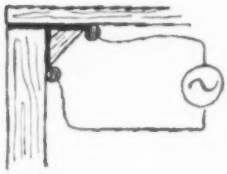
Curved laminated sections are being increasingly used in contemporary furniture, and with dielectric heating rapid production can be achieved using wooden shaping blocks in single daylight presses. An alternative method of providing heat by conduction from heated metal strips becomes increasingly slower as the total section thickness rises above 0.05 inch, as shown below:



TOTAL THICKNESS OF LAMINATIONS	*Comparative heating times in minutes*	
	CONDUCTION	DIELECTRIC
1 inch	20	4
$\frac{3}{8}$ inch	$5\frac{1}{2}$	$2\frac{1}{2}$
0.6 mm veneer	1	$1\frac{1}{4}$

Furniture Assembly

Because of the savings in glueing processes already instanced, dielectric heating is being extensively used in the furniture trade. It leads also to reductions in labour and floor space, with the elimination of assembly jigs. The heating equipment can be placed directly in the production line, cutting handling to a minimum.



Resin-bonded Wood Chipboard

A substitute for natural timber is made from wood waste and chippings, broken down to a coarse size, mixed with synthetic resin and heated under pressure. Dielectric heating gives quick and uniform heating, and increased fluidity reduces the power required for the final pressing and curing operation. In a continuous process, the length of the press required is also reduced.

Blockboard

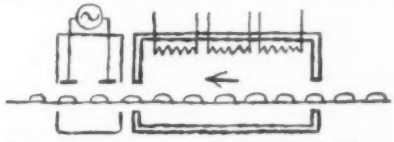
Production of blockboard by edge glueing strips of wood also provides an excellent use for dielectric heating, since considerable savings in time and labour can be effected owing to the large areas of glue line involved.

Other Resin-bonded Products

Dielectric heating is also used in the manufacture of other resin-bonded or impregnated products such as grinding wheels, impregnated woods, fabrics, felts, glass-fibre and similar products.

Foodstuffs

Increased use is being made of dielectric heating in many processes connected with foods; these include de-freezing and melting, sterilisation and disinfestation, drying of breakfast cereals, dog biscuits, rusks etc., heating of nuts to facilitate shelling and other similar types of application. Although some cooking processes are technically



possible, as for example bread baking, the 'unbrowned' product has so far proved unacceptable to the public and a completely dielectric process uneconomical. When combined with conventional baking, however, as now in the biscuit trade, where dielectric heating is being used to complete the baking of biscuits, it can produce normal biscuits in $\frac{1}{3}$ to $\frac{1}{2}$ the usual baking time.

There are in fact so many potential applications of dielectric heating (and these applications are increasing daily as the chemical industry develops new products, as for example synthetic fibres) that the selection given in the present series of data sheets covers only a part of the whole field.

For further information get in touch with your Electricity Board or write direct to the Electrical Development Association, 2 Savoy Hill, London, W.C.2. Telephone: TEMple Bar 9434.

Excellent reference books on electricity and productivity (8 6 each, or 9 - post free) are available—"Induction and Dielectric Heating" is an example.

E.D.A. also have available on free loan in the United Kingdom a series of films on the industrial uses of electricity. Ask for a catalogue.

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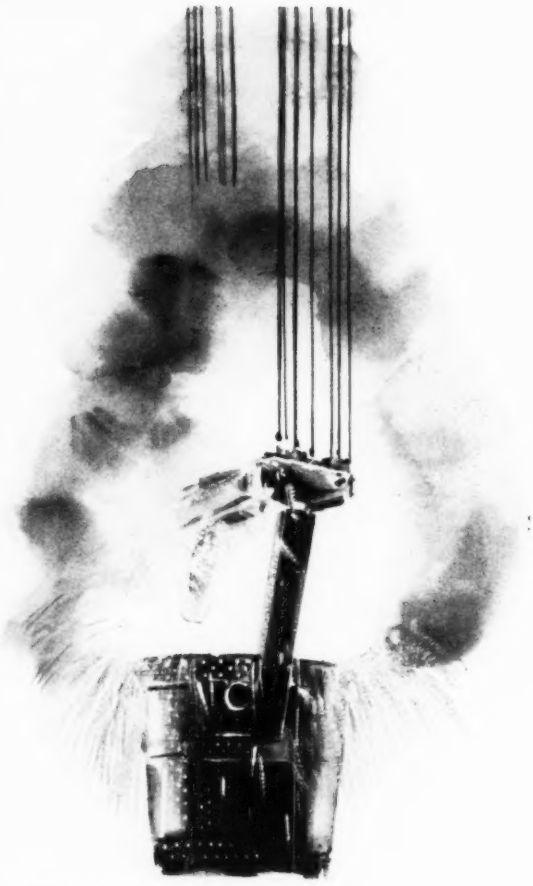

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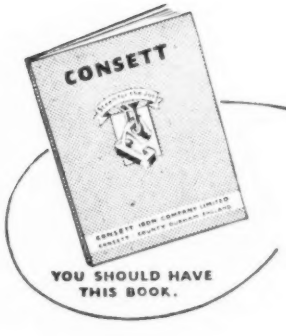

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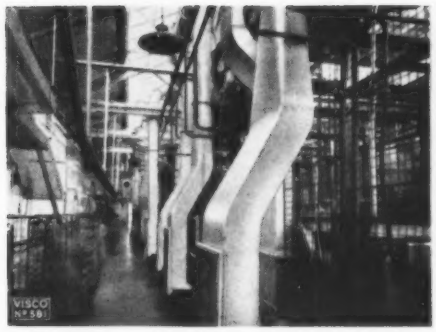

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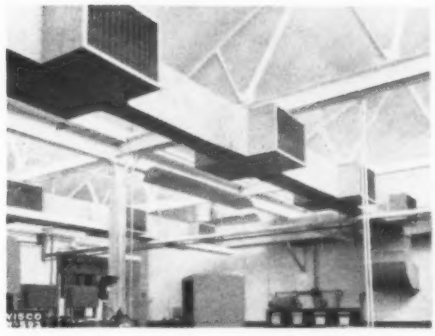

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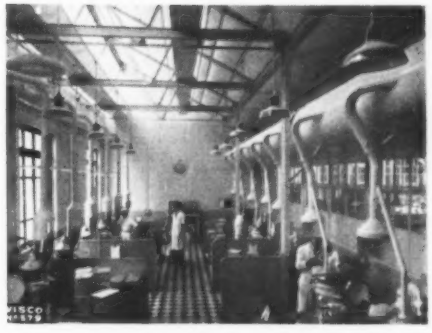

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Sixth Report of the Corrosion Committee

The corrosion of iron and steel costs industry millions of pounds every year, and research on the best ways of saving this money is of the highest importance. The Sixth Report of the Corrosion Committee, which was founded in 1928 and is now administered by BISRA, has recently appeared as No. 66 in the Institute's Special Report series; it contains an account of the researches carried out during the past twenty years.

The Report covers the corrosion of iron and steel in air, in soil, and in water; it details the results of numerous experiments and recommends methods of protection against the many different forms of corrosion. Reports of tests carried out to ascertain the corrosion rates of different kinds of bare irons and steels all over the world, and of steels protected with paint and/or metal coatings, are given.

Two other interesting researches described in the Report concern the corrosion and protection of steel at elevated temperatures up to 350°C, and marine corrosion. As a result of the latter, several reliable anti-corrosive compositions have been formulated, some of which are in use on H.M. ships and the "Queens".

The Report, which consists of 228 octavo pages, illustrated with 20 line and 21 half-tone figures, and bound in cloth with stiff board covers, costs £3 3*s*. 0*d*. (postage 1*s*. 0*d*.), and may be obtained from:

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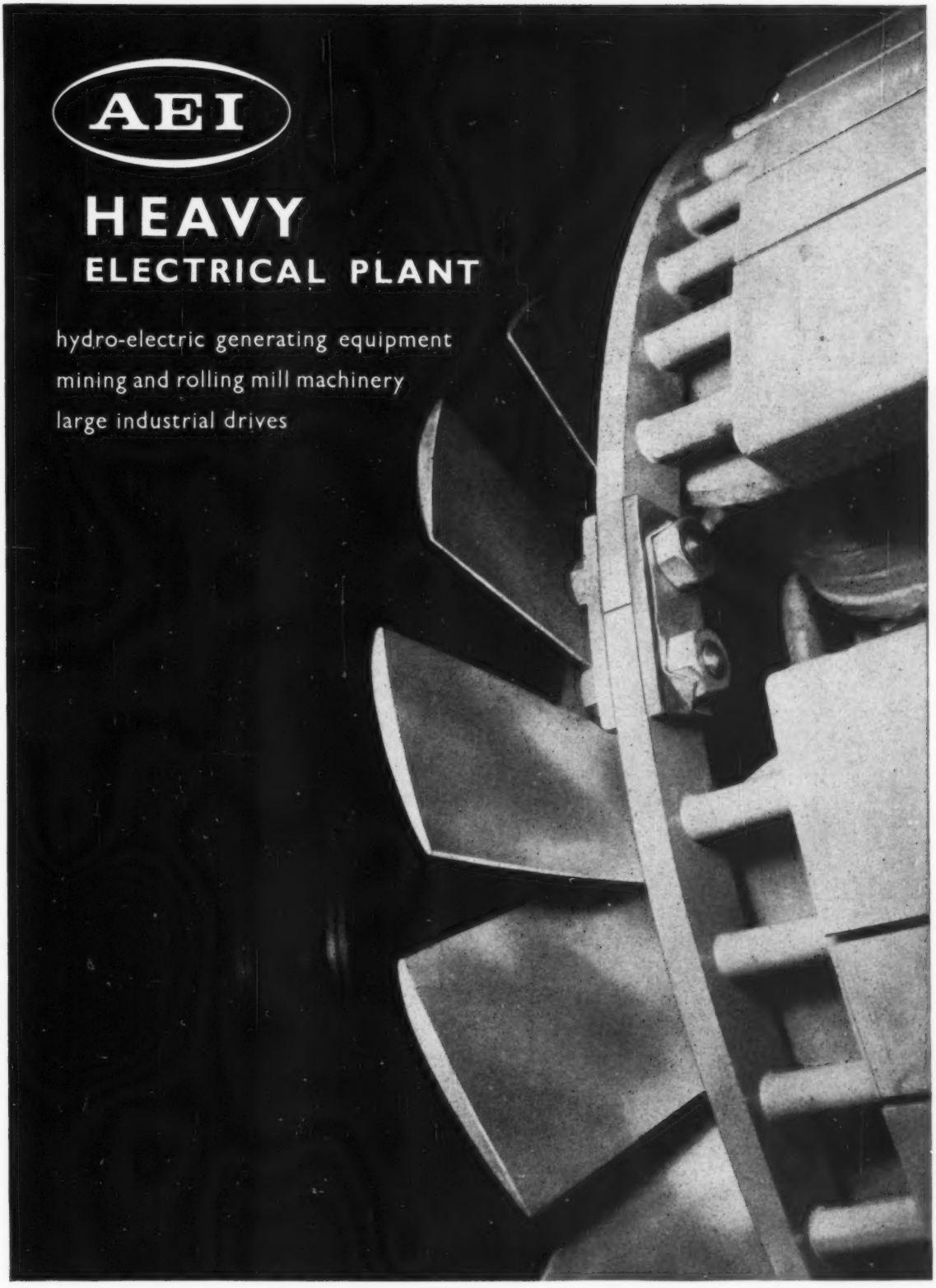

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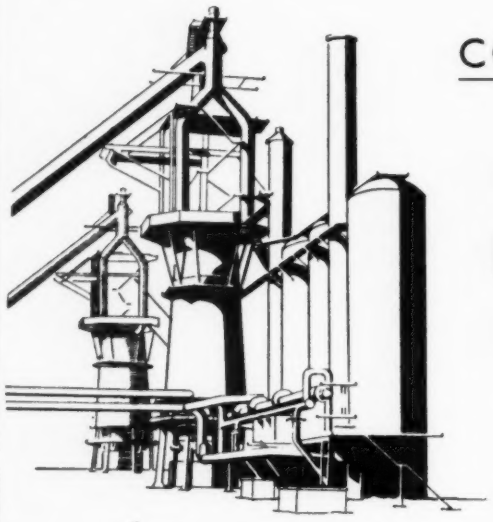

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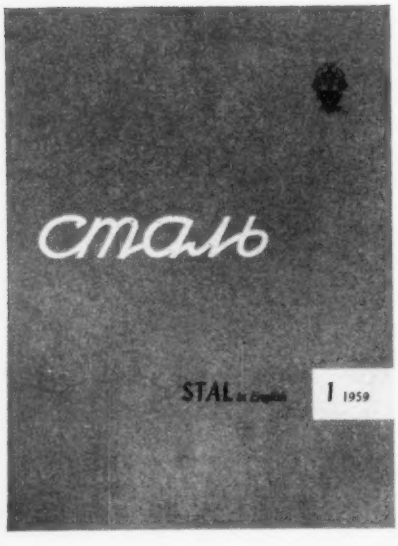

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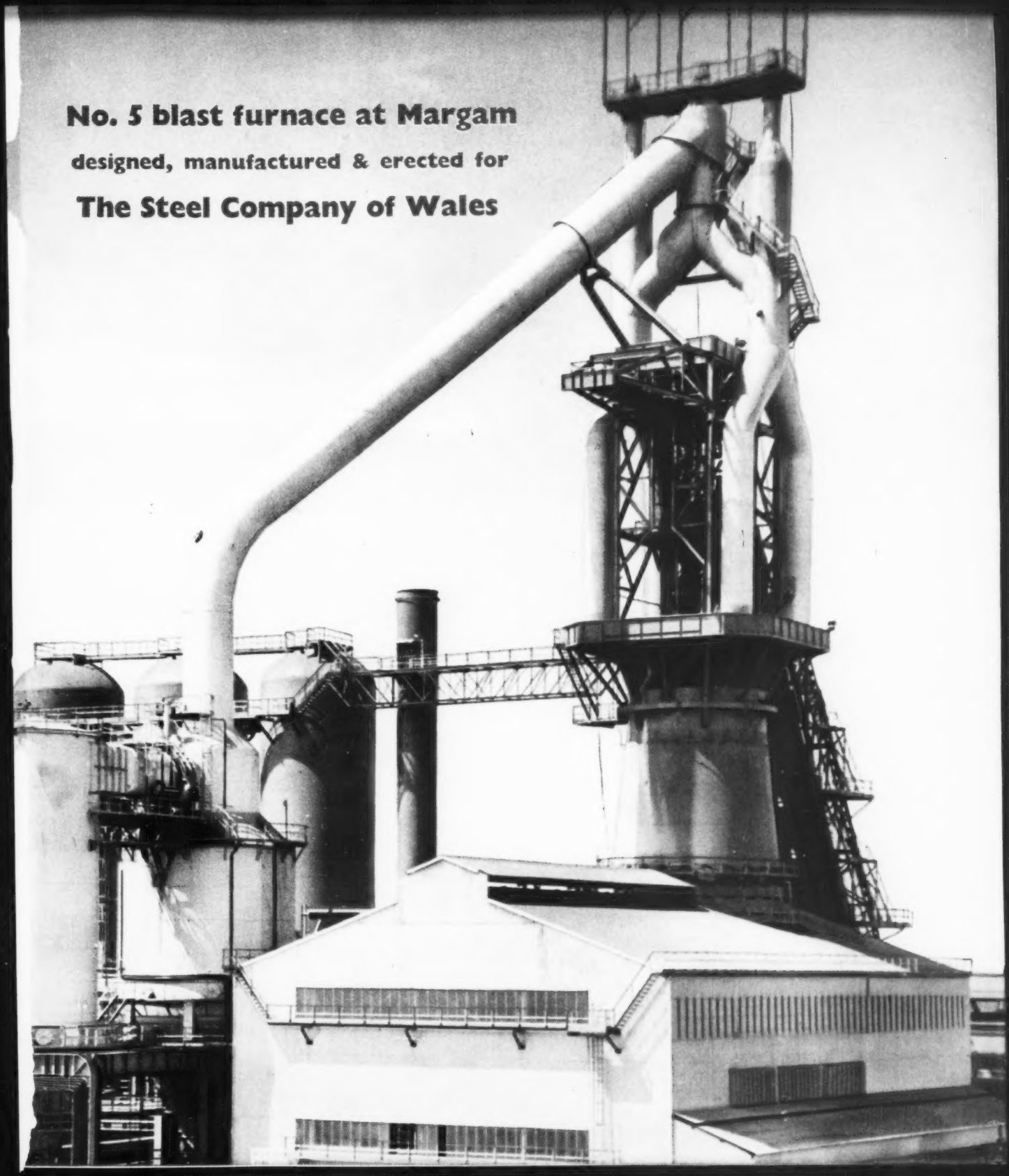

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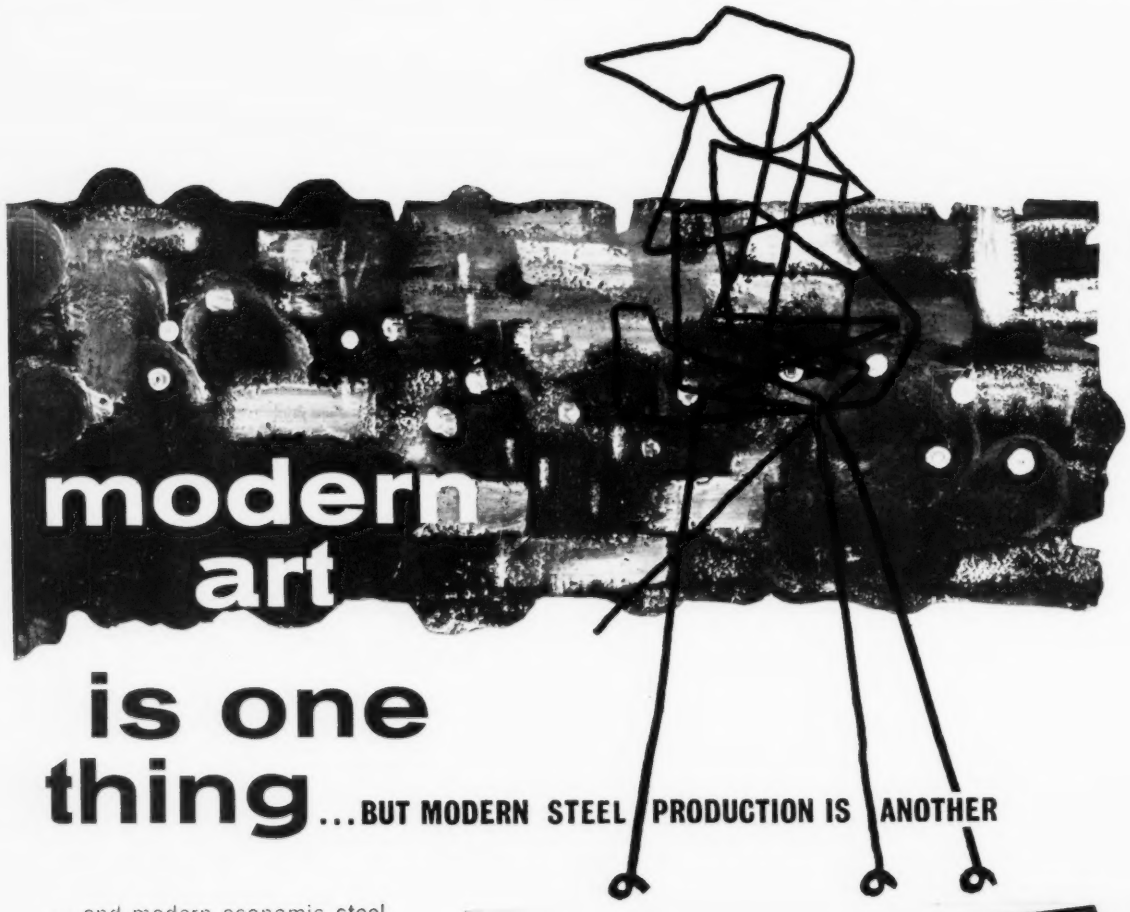

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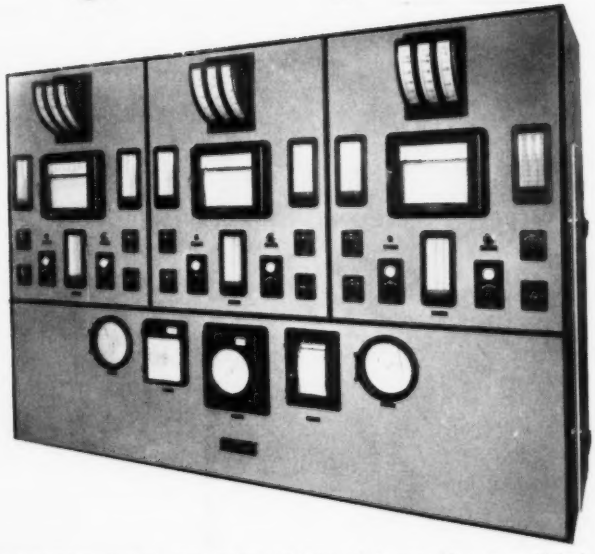

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